JANUARY 1959

# JOURNAL OF THE IRON AND STEEL INSTITUTE

4 GROSVENOR GARDENS · LONDON SWI

*Established 1869 · Incorporated by Royal Charter 1899*

VOL. 191 PART 1

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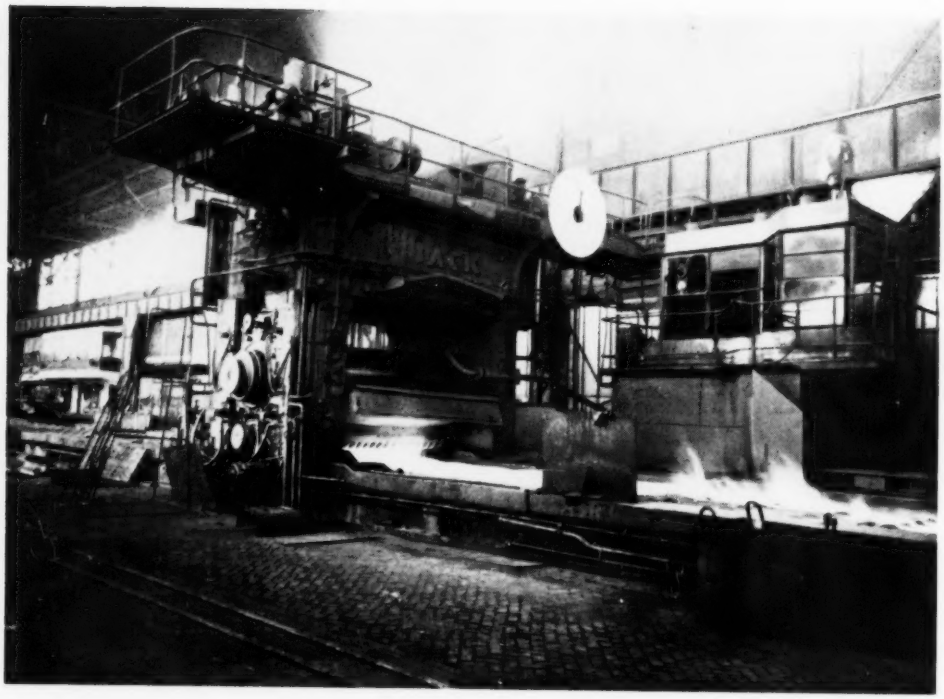

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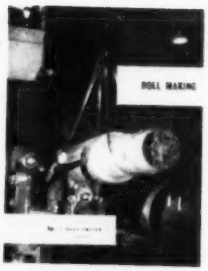
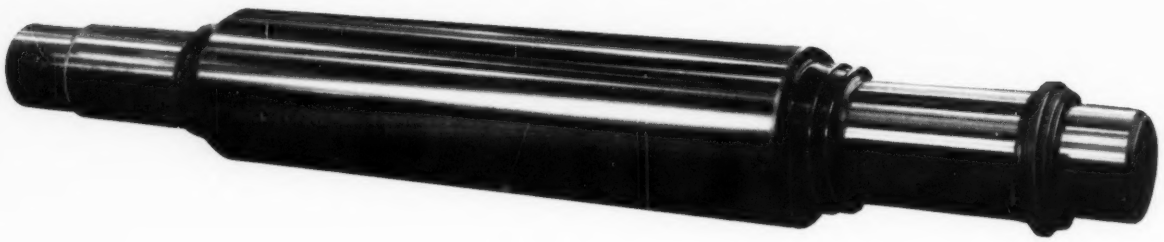
The U.X. Alloy Indefinite Chill Roll offers the maximum resistance to firecracking and spalling in the rolling of PLATES. U.X. has been outstandingly successful when used for work rolls in 4-High Reversing Plate Mills and also in the earlier stands of 4-High Continuous Hot Strip Mills. The alloys present ensure that the rolls wear with the minimum fall-off in surface hardness until the working life is completed.




*The U.X. Alloy Roll is described fully in our brochure "ROLL MAKING," which shows this and many other types of roll manufactured by DAVY-UNITED.*



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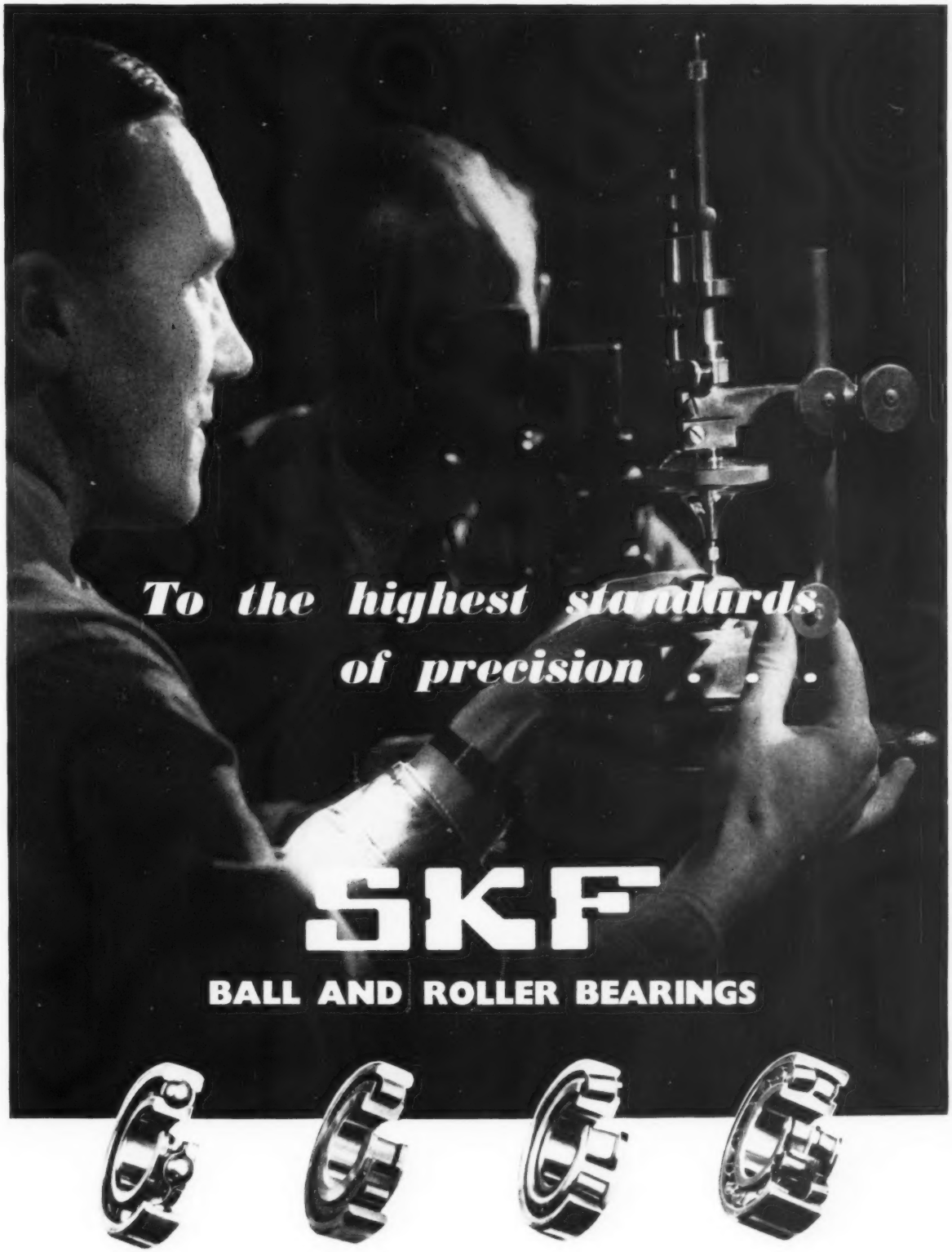


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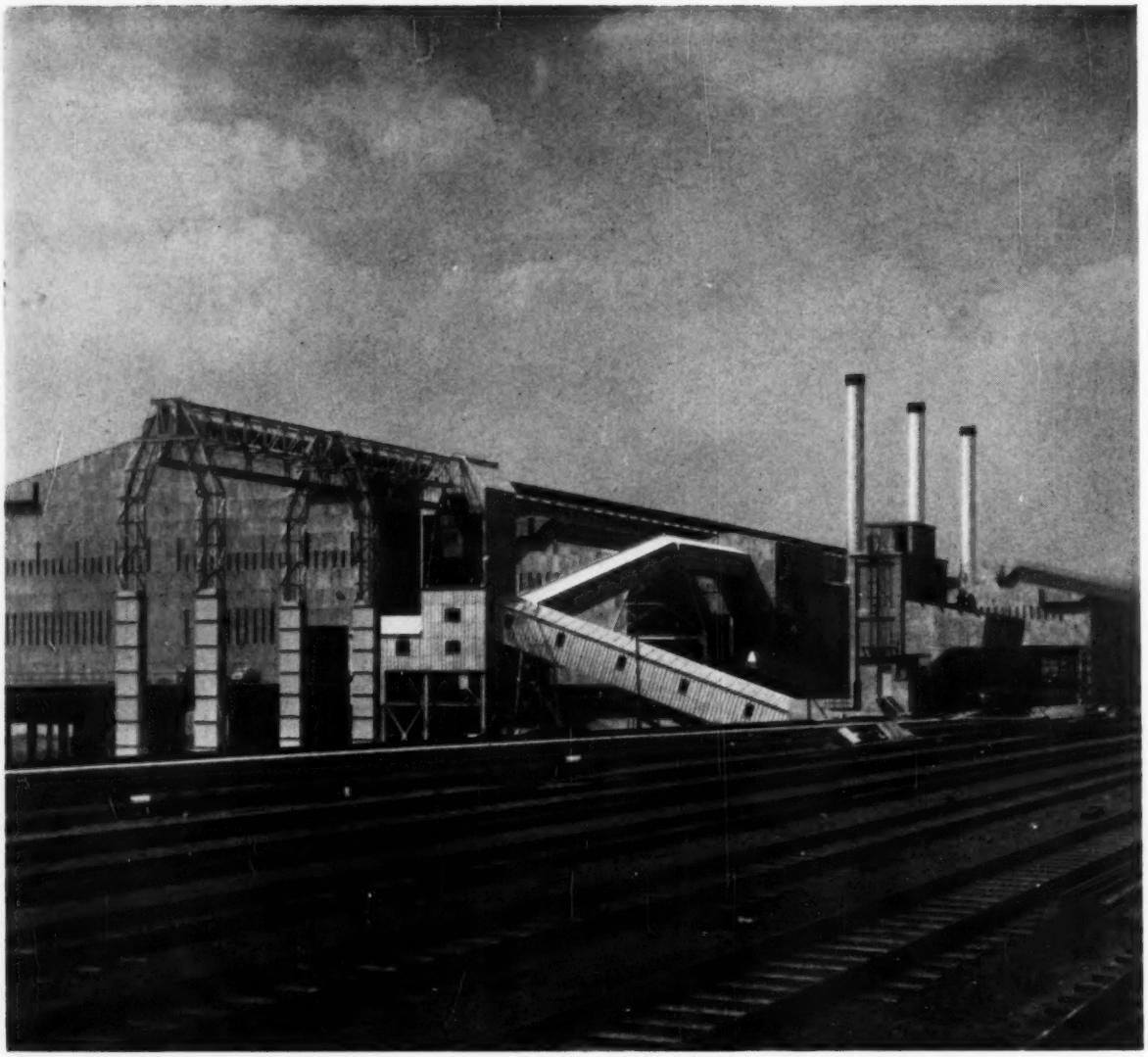

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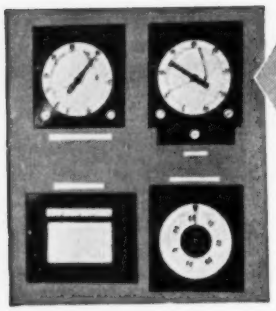

## Honeywell **Area** Flow-meter

Colville's Dalzell Works. Area Meter Body on creosote-pitch line in valve cabin transmits to ElectroniK Inductance Bridge Receiver on control panel, top right.

## ...Simple...Accurate...Rugged

Colville's Dalzell works are among many which now recommend the Honeywell Area Flow Meter and ElectroniK Inductance Bridge Receiver as the *simplest* means of metering heavy oil, tar and creosote-pitch. Features of the system include:

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- **Fully automatic control . . . pneumatic . . . electric**
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THESE GIVE FULL DETAILS
Specification Sheet 226a (Area Meter Body)
and Special Bulletin HB 1540
(ElectroniK Inductance Bridge Receiver).
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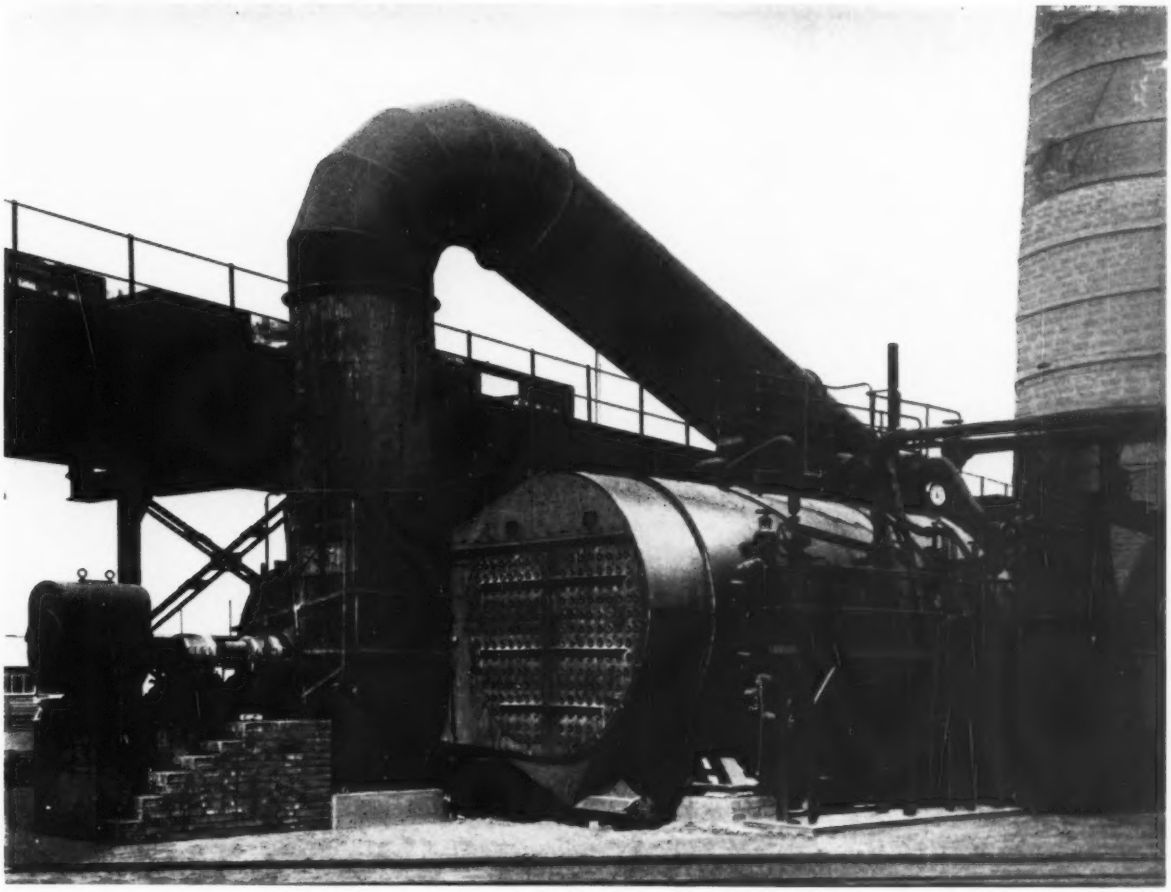
# WASTE HEAT RECOVERY IN THE STEEL INDUSTRY



*Photograph by courtesy of Messrs. John Lysaght's Scunthorpe Works Ltd.*

One of a battery of SPENCER-BONECOURT Waste Heat Boilers operating on Open Hearth Steel Melting Furnaces at John Lysaght's Scunthorpe Works Ltd., Scunthorpe.

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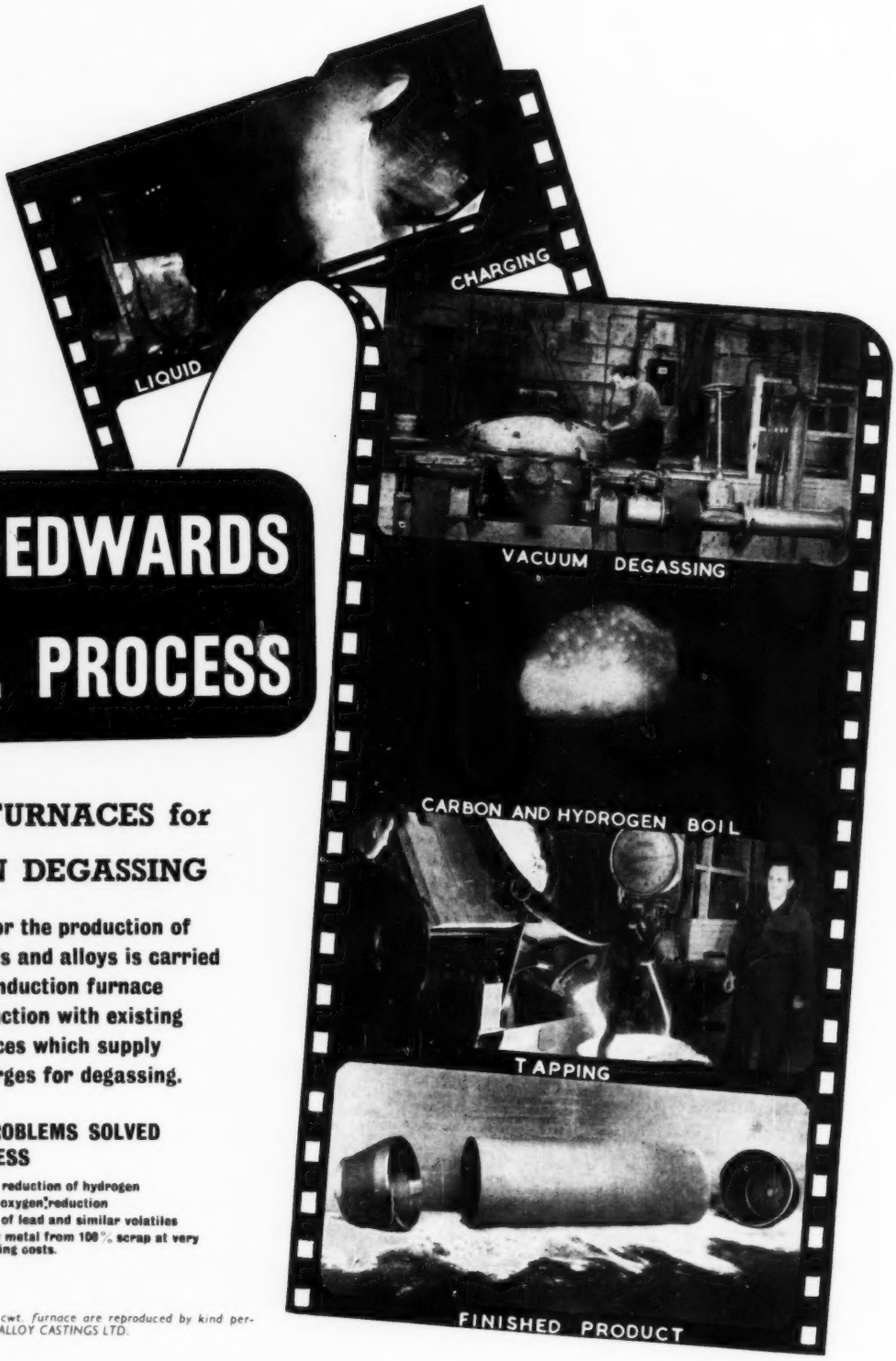


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- High quality metal from 100% scrap at very low processing costs.

*The illustrations of a 15 cwt. furnace are reproduced by kind permission of SHEEPBRIDGE ALLOY CASTINGS LTD.*

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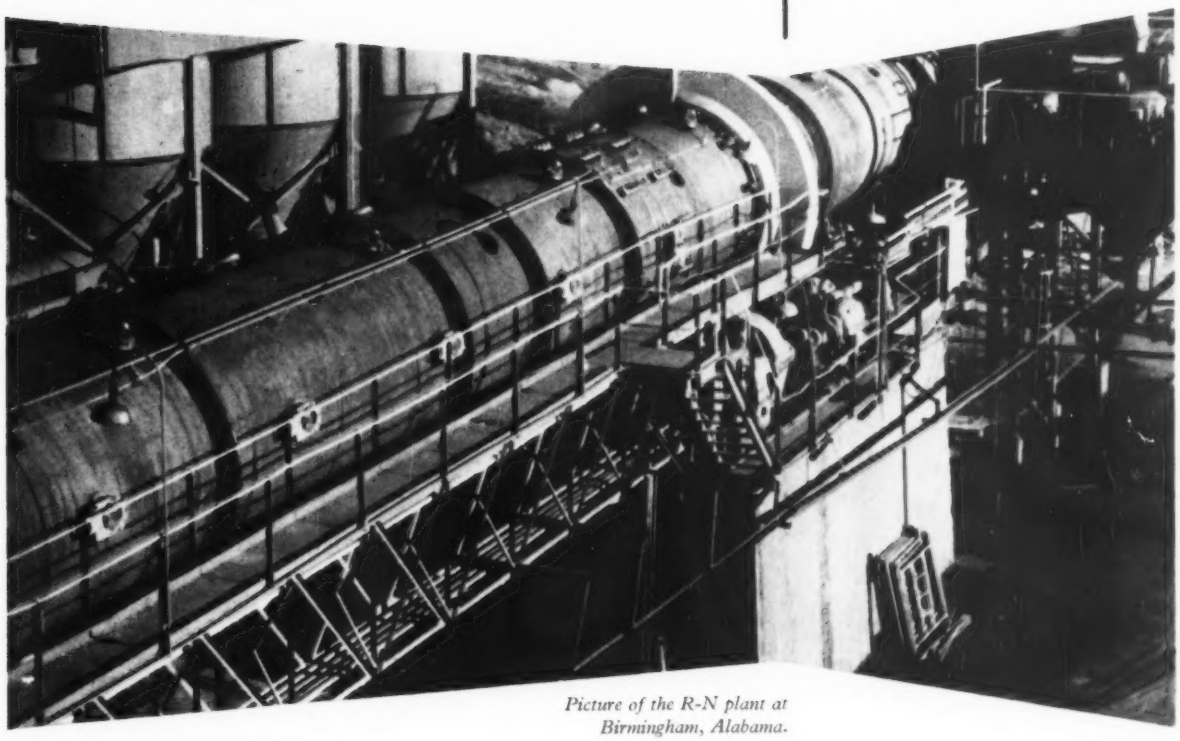
# THE R-N PROCESS

for the direct reduction of iron ore



*Picture of the R-N plant at Birmingham, Alabama.*

## R-N Product replaces pig iron as well as scrap

Within the E.C.S.C. countries an increase in steel production from 60 mill. tons in 1957 to 85 mill. tons in 1965 has been estimated. A similar percentagewise growth is foreseen for Great Britain.

In many countries this increase in production will bring difficulties in regard to sufficient supply of iron ore, coking coal and iron scrap.

The R-N process provides the world's steel industry with a new and inexpensive method for reduction in a rotary kiln of a wider variety of iron ores utilizing various types of solid, low-cost carbon fuels.

The capital cost of an R-N plant is about half that required for a blast furnace complex of corresponding size, and the production cost for R-N metallic iron briquettes will be considerably lower than that for pig iron.

Low stable price and uniform composition make R-N iron an excellent and competitive substitute for scrap.

The development of the R-N process to a commercial process is the result of collaboration between the Republic Steel Corporation and the National Lean Company through the R-N Corporation, New York.

*Representative for Europe:*

**R-N PROCESS AGENCY**

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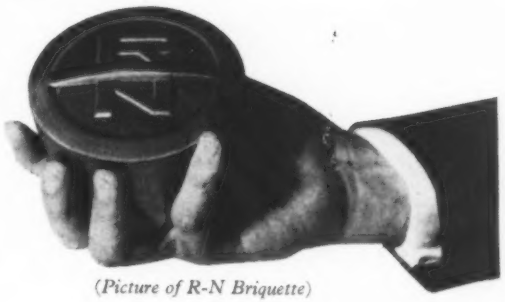

*(Picture of R-N Briquette)*

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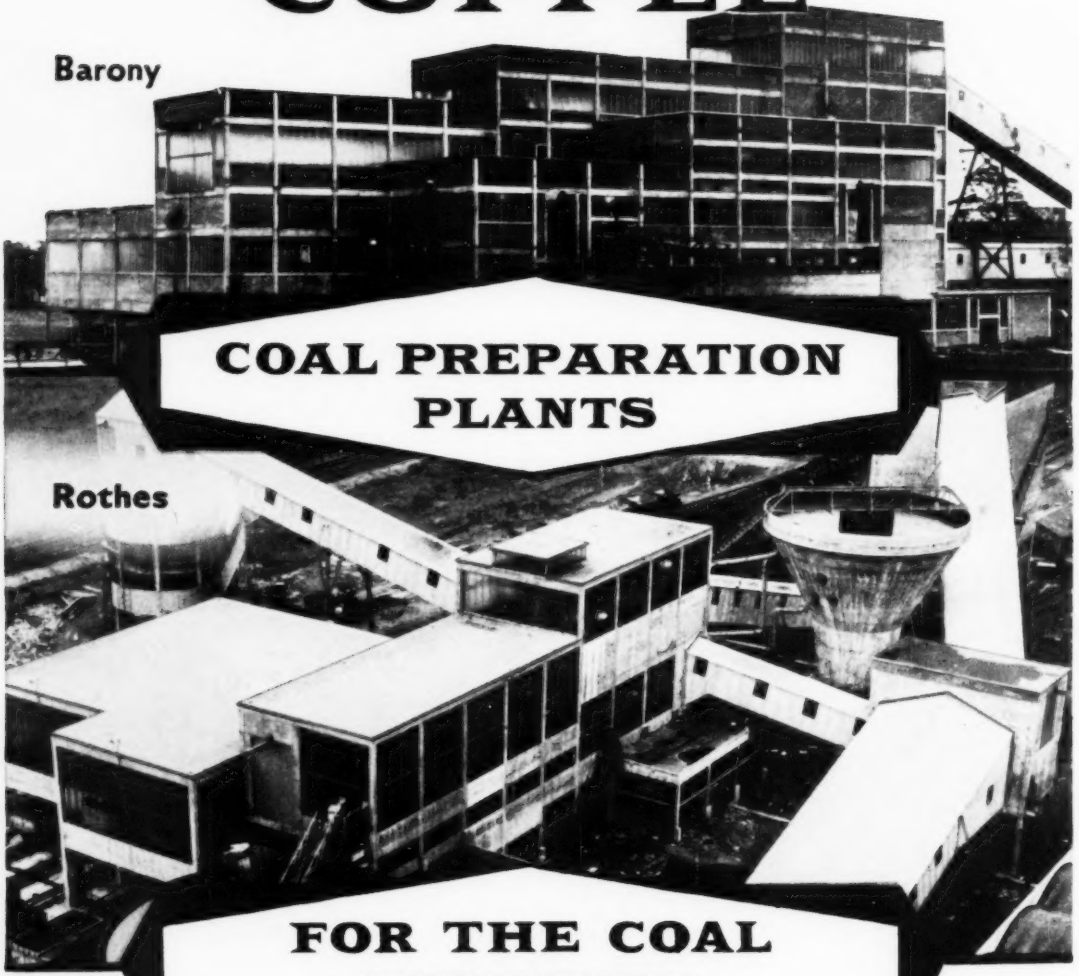

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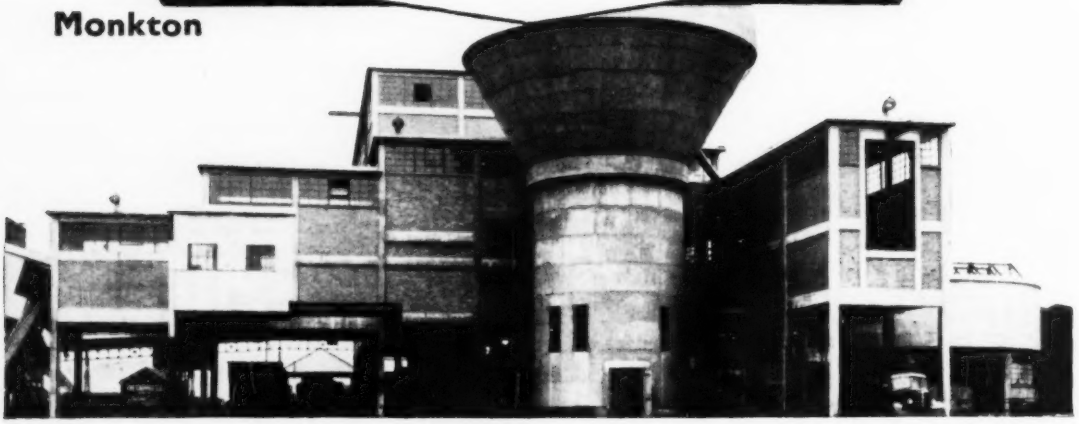

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*(Main Contractors : Messrs. Constructors John Brown)*

For their new 'ACRILAN' factory at Coleraine, N. Ireland, Messrs. Chemstrand have installed a 10,000 galls. hr. 'Boby' demineralisation plant for process and boiler feed water.

*The plant comprises the following sections :*

Coagulation and Settlement — Filtration — Cation Exchange — Weakly basic Anion Exchange — Degassing — Mixed Bed Ion Exchange.

## 2 VEITSILUOTO O/Y, Finland

The Veitsiluoto paper mills of Finland have installed a 13,400 galls/hr. 'Boby' demineralisation water treatment plant for feed water for boiler working at 1,750 p.s.i.

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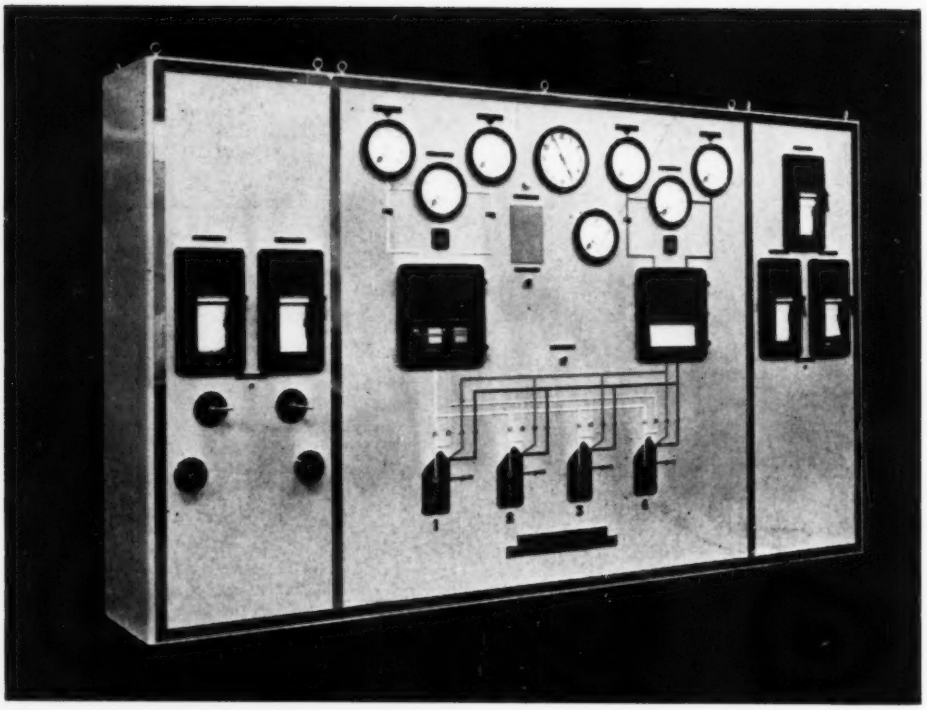

# Another successful "first"

The first automatic electronic control installation
for a Bessemer Converter plant at
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Oxygen/Steam Ratio
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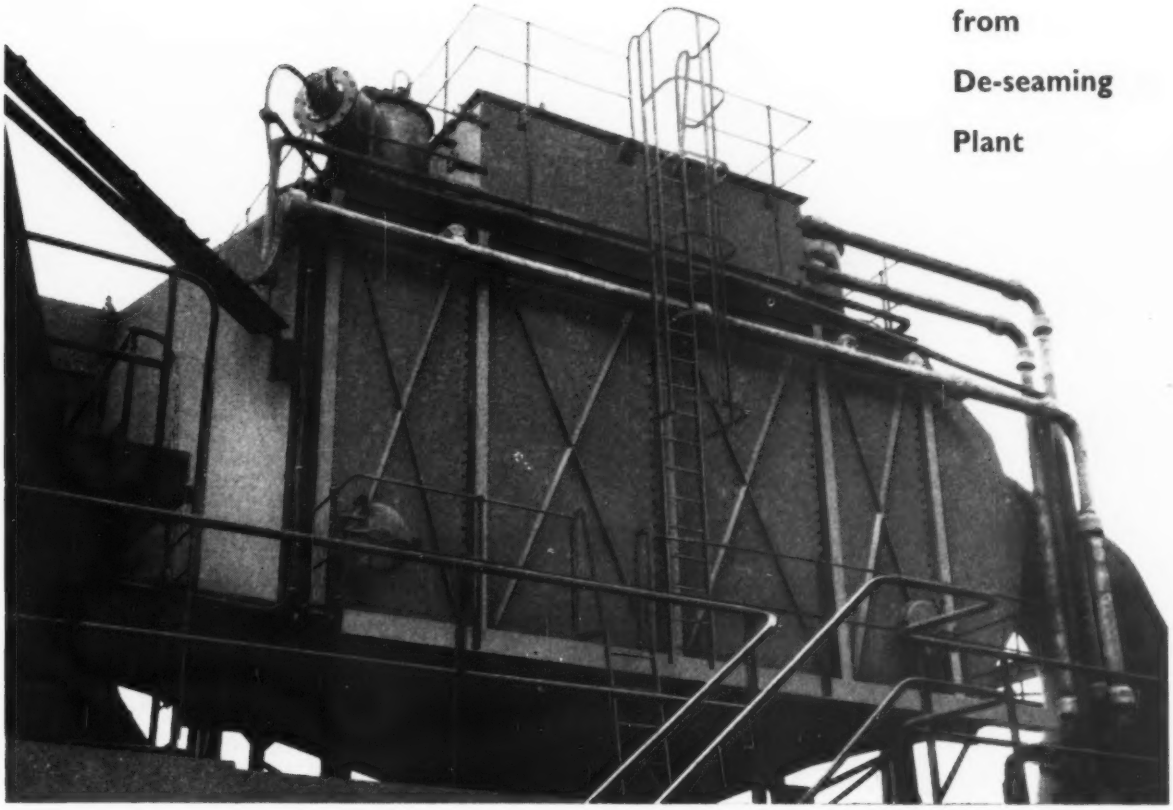
# HOLMES-ELEX ELECTRICAL PRECIPITATORS

**Iron oxide removal from De-seaming Plant**



The Holmes-Elex Electrical Precipitator illustrated above is installed at the Corby Works of Messrs. Stewarts & Lloyds for the purpose of cleaning exhaust air from the collecting hood enclosing the De-seaming Plant. The fine particles associated with iron oxide fume, and which are responsible for the dense brown colour are removed resulting in no visible stack emission.

*For full details please write for Publication No. 69/24*

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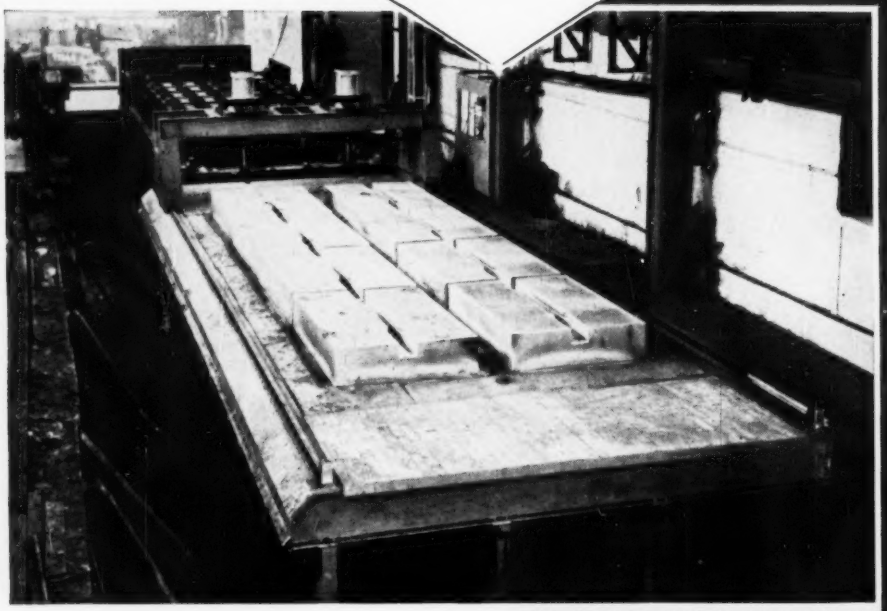
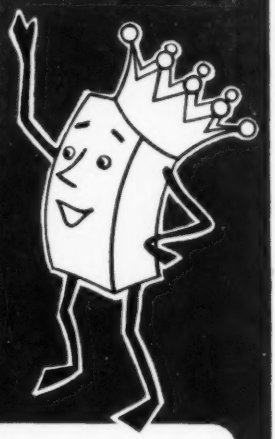


*Photograph reproduced by courtesy of Appleby — Frodingham Steel Company (Branch of The United Steel Companies, Ltd.) and Stein and Atkinson, Ltd.*



Electric soaking pit, designed and built by Stein and Atkinson Ltd., for the Appleby—Frodingham Steel Company (Branch of the United Steel Companies Ltd.), Scunthorpe

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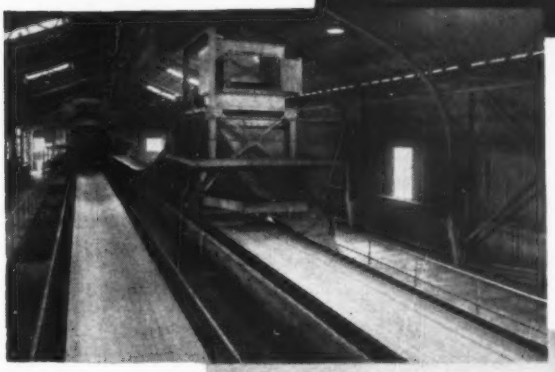

Trippers feeding Ore to Dockside Bunkers.

Boomstacker stock-piling Iron Ore.

Iron Ore Stocking out Bridge.

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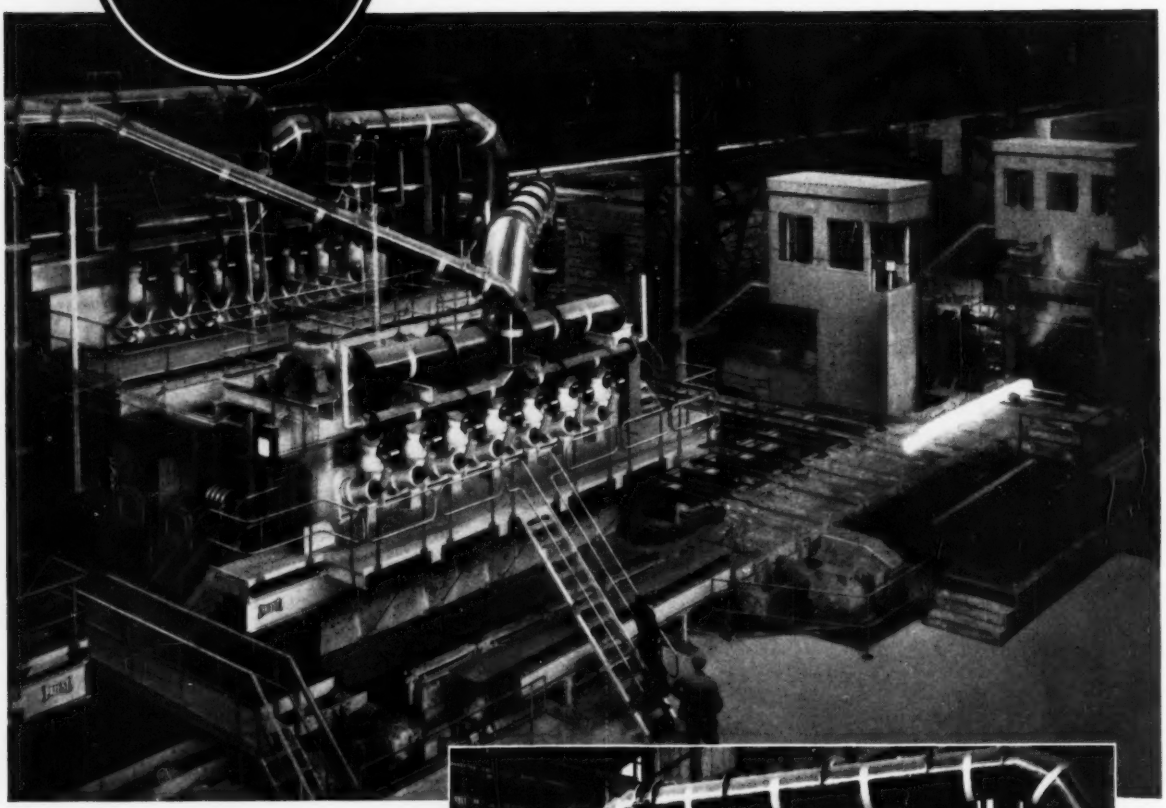

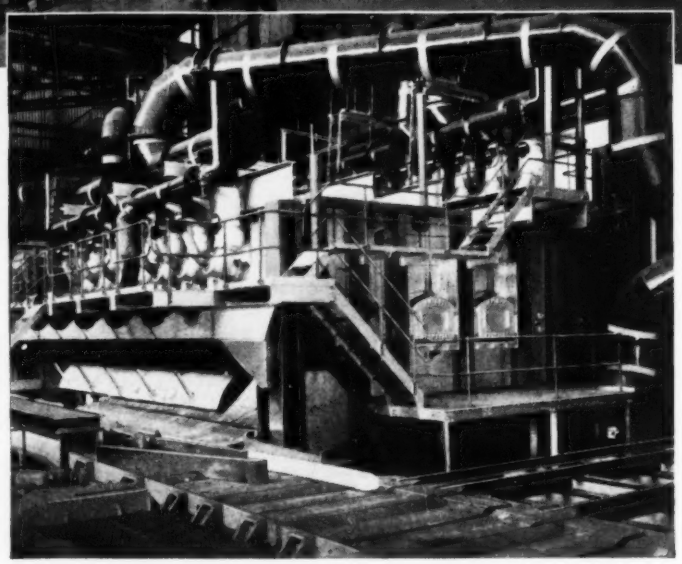

The photographs show one of two Priest 3-Zone Continuous type gas fired recuperative Bloom Re-heating Furnaces for Medium Section Mill. Installed at the Cleveland Works of Dorman Long (Steel) Limited.

We also supply similar Furnaces for re-heating slabs or blooms suitable for gas or liquid fuel firing.

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BIGWOOD

The complete Bigwood slitting line shown above comprises coil loading bogie, pay-off reel, oiling machine, gang slitter, re-coiler with hydraulic push-off and coil stripper carriage.

The line can deal with strip up to a maximum width of 20 inches in thicknesses from 0.04 to 0.64 ins. at a pass speed of 250/450 ft. per minute.

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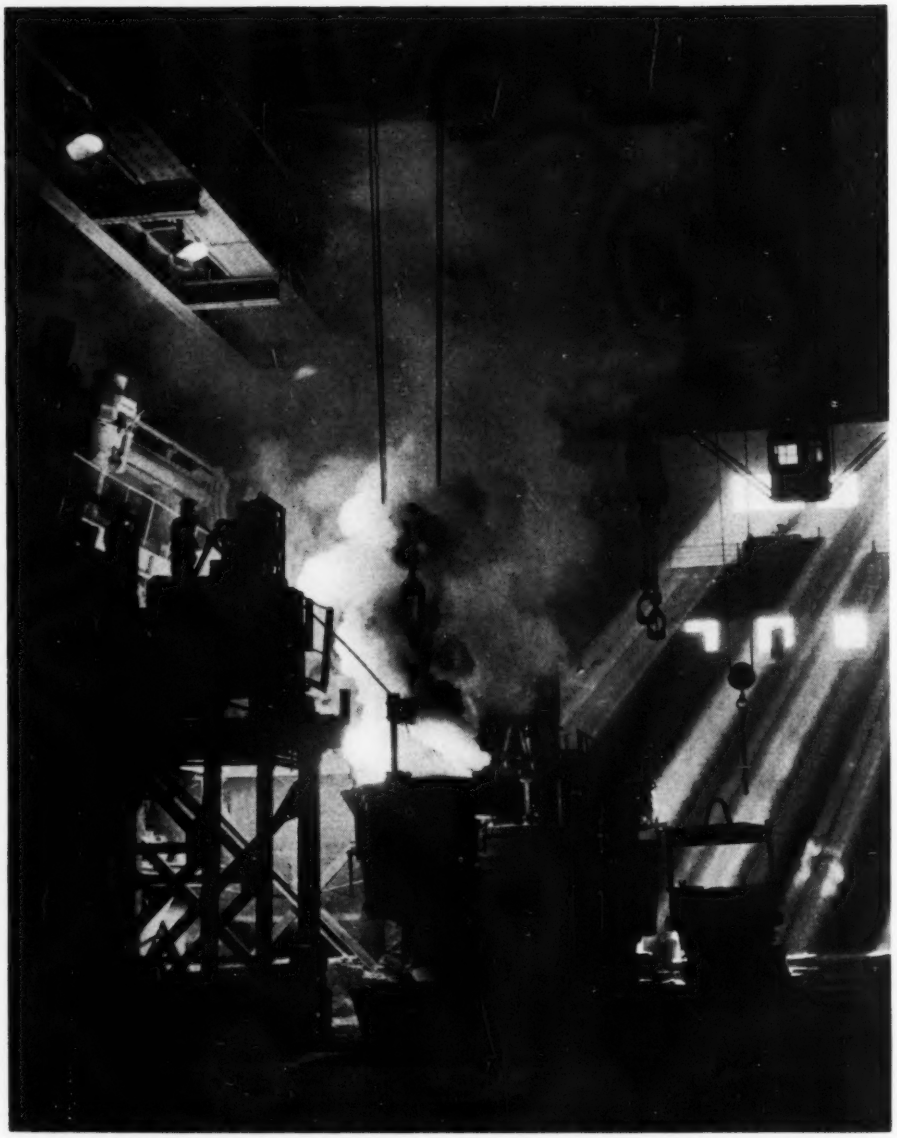
# Industry demands Steel —and Allen West Control Gear



Furnace tapping in the Electric Arc Melting Shop of Messrs Thos Firth & John Brown Ltd Sheffield, by whose courtesy this photograph is reproduced

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by S & A

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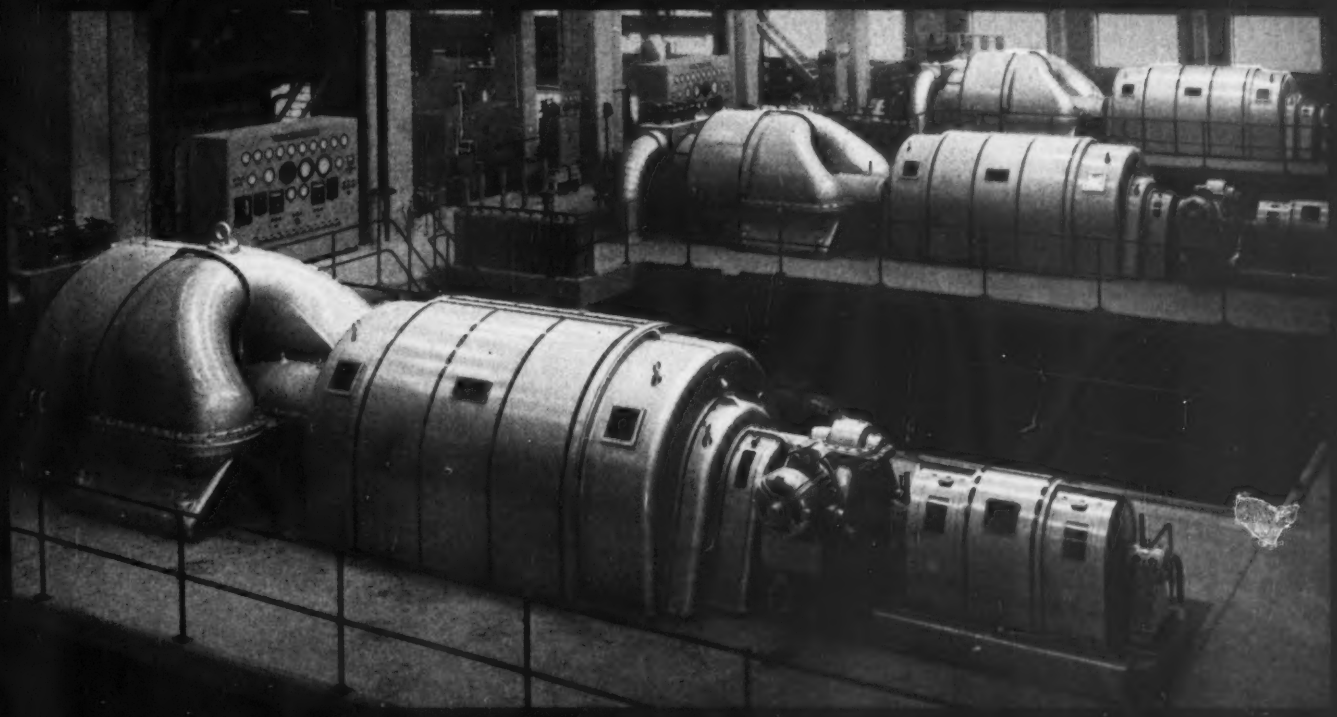
# steel

Ravenscraig, near Motherwell in Scotland, is the site of the latest and most important addition to the iron and steel industry of Scotland to which Colvilles Ltd. have already contributed so much.

Here, from an undeveloped site, the first phase of this expansion, contributing more than a thousand tons of iron daily, was put into operation within three years. In this achievement the G.E.C. assumed responsibility for the whole of the power plant. This included the civil work, cooling towers, boilers, three turbo alternators, two turbine driven blast furnace blowers, two turbine driven pumps, and all associated switchgear, transformers and ancillary equipment to control the gas, steam and electric supplies.

Electrical connection to the South of Scotland Electricity Board system and to other works of Colvilles Ltd. was also undertaken by the G.E.C.



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G.E.C. generators and blowers

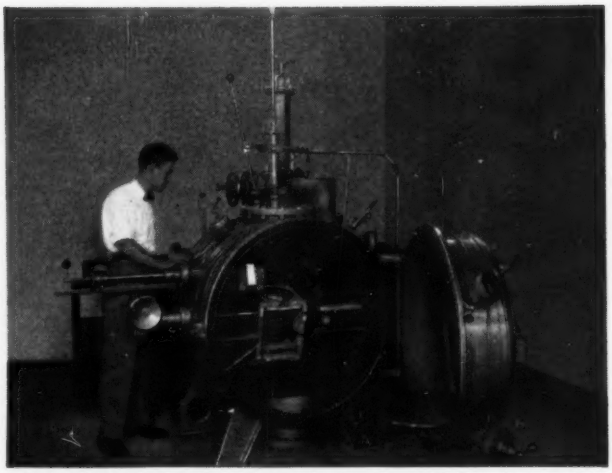
THE GENERAL ELECTRIC COMPANY LIMITED OF ENGLAND · HEAD OFFICE: MAGNET HOUSE, KINGSWAY, LONDON, W.C.2

WILD BARFIELD NRC*

# Vacuum Melting

## INDUCTION AND ARC FURNACES

The experience of the National Research Corporation, who have built and operated more high vacuum furnaces than any other company in the world, and the facilities in Britain of Wild-Barfield combine to offer unsurpassed vacuum melting equipment.
Both arc and induction furnaces are available, the former with consumable or non-consumable electrodes, cold mould or skull type. Induction heated equipments include the N R C patented semi-continuous vacuum melting furnaces.



*Model 2555 Vacuum Induction Furnace with melting capacity of 50 pounds of steel. Other standard furnaces have capacities of 12 to 3,000 pounds.*

### *INDUCTION HEATED FURNACES FOR ...*

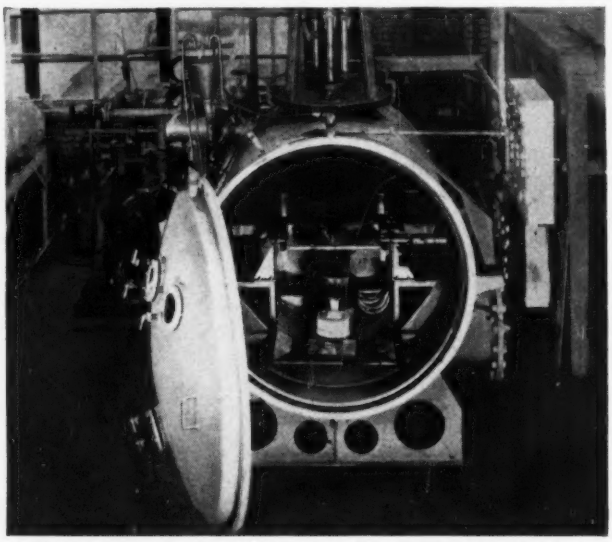
Melting, Alloying, and Purifying; Centrifugal, Ingot, Shape, and Investment Casting; Sintering, Brazing, Degassing and Heat-Treating.

### *ARC FURNACES FOR ...*

Melting and alloying high temperature or reactive metals such as niobium, tantalum, cobalt, zirconium and titanium.

### OTHER VACUUM EQUIPMENTS

- **Vacuum Fusion Gas Analyser**
- **Hydrogen Analyser**
- **Resistance Furnaces**
- **High Vacuum Diffusion Pumps**



*Model 2705 Non-Consumable Arc Skull Furnace with a capacity of 50 pounds of titanium. Other standard vacuum arc furnaces have capacities of 8 to 10,000 pounds of titanium.*

* NRC is the trade-mark of the National Research Corporation, registered in the United States Patent Office.

ELECTRIC WILD BARFIELD FURNACES

***FOR ALL HEAT-TREATMENT PURPOSES***

**WILD-BARFIELD ELECTRIC FURNACES LIMITED**

ELECFURN WORKS, OTTERSPOOL WAY, WATFORD BY-PASS, WATFORD, HERTS. Telephone: Watford 06091 (8 lines)

NRC 2

The above illustration shows a general view of the annealing furnaces, installed for the Steel Company of Wales at Velindre Works. The illustration on the right shows the H.N. Atmosphere Gas Plant also installed by Salem at Velindre Works.

**DIRECT FIRED COIL ANNEALING FURNACES**

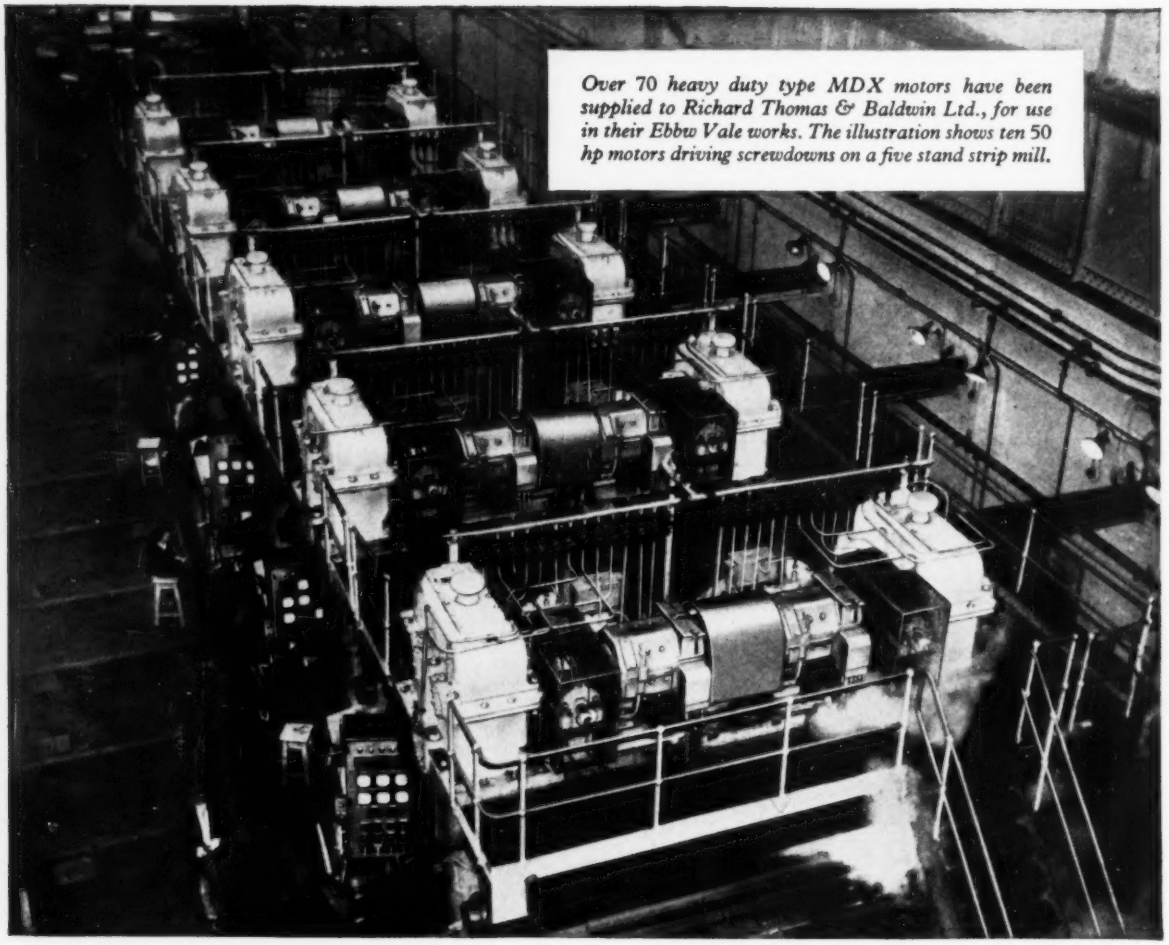
Salem design, build and put into operation complete plant with all ancillaries.

**Salem**

**SALEM ENGINEERING COMPANY LIMITED**

MILFORD HOUSE MILFORD DERBY

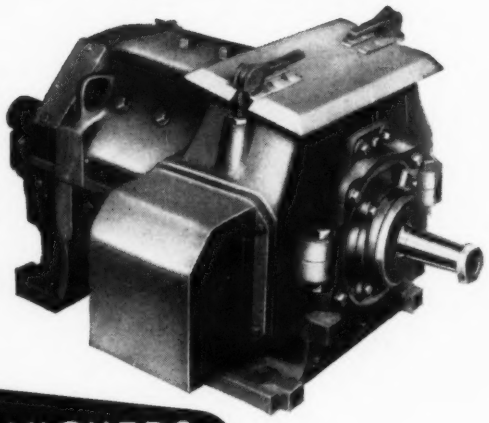
Telephone Duffield 2271 Telegrams Kosmar "Belper"



*Over 70 heavy duty type MDX motors have been supplied to Richard Thomas & Baldwin Ltd., for use in their Ebbw Vale works. The illustration shows ten 50 hp motors driving screwdowns on a five stand strip mill.*

# All motors here are type MDX

Metrovick type MDX motors have been specially designed to operate under the arduous conditions of iron and steel works. Their strength and reliability, and the exceptionally low inertia of the rotating parts make these machines ideal for driving a wide range of steelworks equipment. The use of mica and glass insulation suitable for high temperature operation enables the motor to withstand heavy overloads.
Available totally enclosed, up to 200 hp, or self-ventilated, up to 250 hp.



METROPOLITAN-VICKERS
ELECTRICAL CO LTD TRAFFORD PARK MANCHESTER, 17

*An A.E.I. Company*

## Metrovick Motors for Trouble Free Service

J 5702

# ARC MELTING SUPREMACY

**Birlec-Efco (Melting) Limited has been formed by Birlec Limited and Efco Limited for the design and supply of all types of electric melting furnaces, including vacuum furnaces, for the ferrous and non-ferrous metals industries, together with smelting furnaces and induction heating equipment. During the last 50 years the parent companies have installed throughout the world:—**

**over 600 ARC FURNACES**

**—combined ratings exceed 1,000,000 kVA**

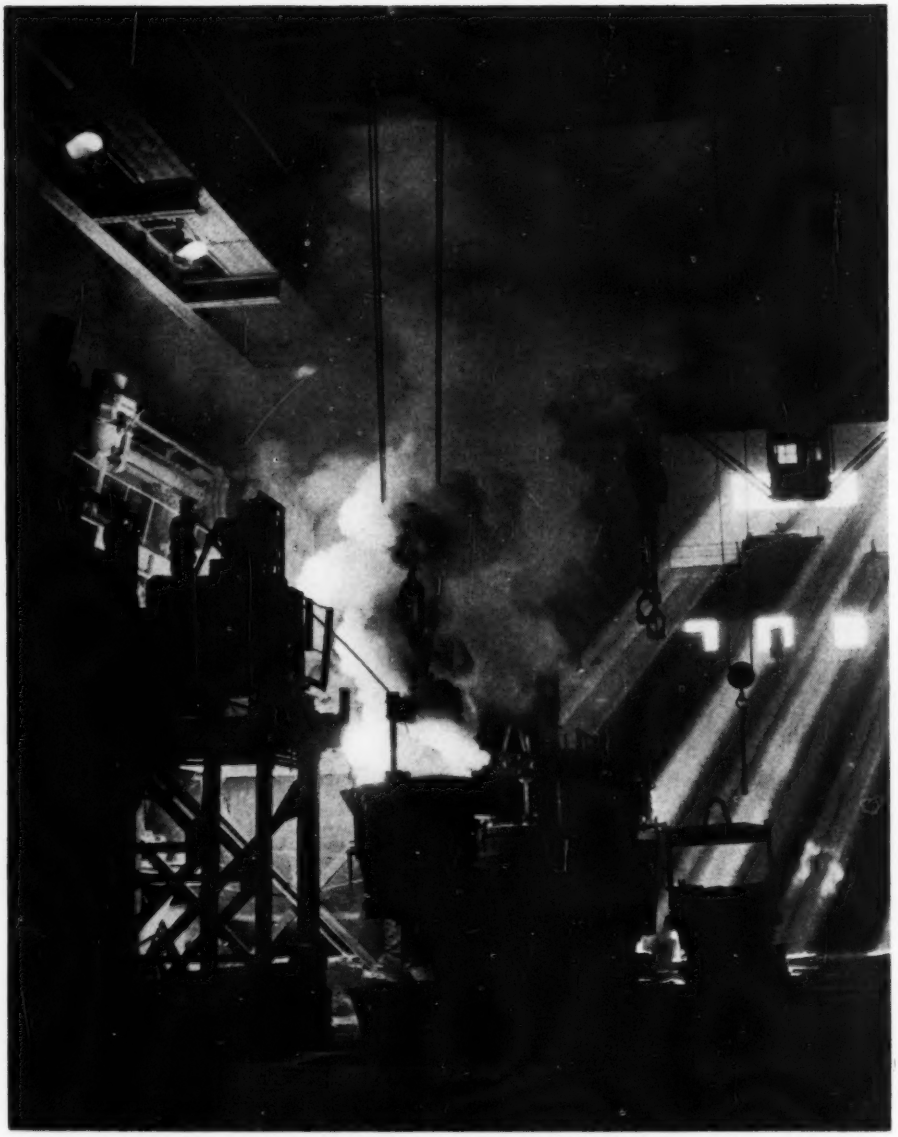
***The directors and staff of the new company are, without exception, drawn from the parent companies. A unique combination of design, experience and skill is thus made available to the metallurgical and engineering industries.***

**BIRLEC-EFCO (MELTING) LIMITED**
WESTGATE · ALDRIDGE · STAFFORDSHIRE
TELEPHONE: ALDRIDGE 52071

BIRLEFCO

SM/BE 4410

# Industry demands Steel —and Allen West Control Gear



Furnace tapping in the Electric Arc Melting Shop of Messrs Thos Firth & John Brown Ltd Sheffield, by whose courtesy this photograph is reproduced

*Whatever the application, for every motor there is an Allen West starter*

ALLEN WEST

* **Designed to B.S. Specification throughout**
* **Complete range—A.C. or D.C. automatic or hand-operated, from fractional to thousands of horsepower**
* **Single units or composite switchboards**
* **Crane Control Gear**
* **All classes of enclosure, from open-type to fully weatherproof**

**ALLEN WEST & CO LTD** BRIGHTON ENGLAND · Telephone: Brighton 23291 · Telegrams: Control, Brighton
*Engineers and Manufacturers of Electric Motor Control Gear and Switchgear*
SUBSIDIARY COMPANIES IN CANADA, SOUTH AFRICA AND RHODESIA AGENCIES THROUGHOUT THE WORLD

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## Britain's

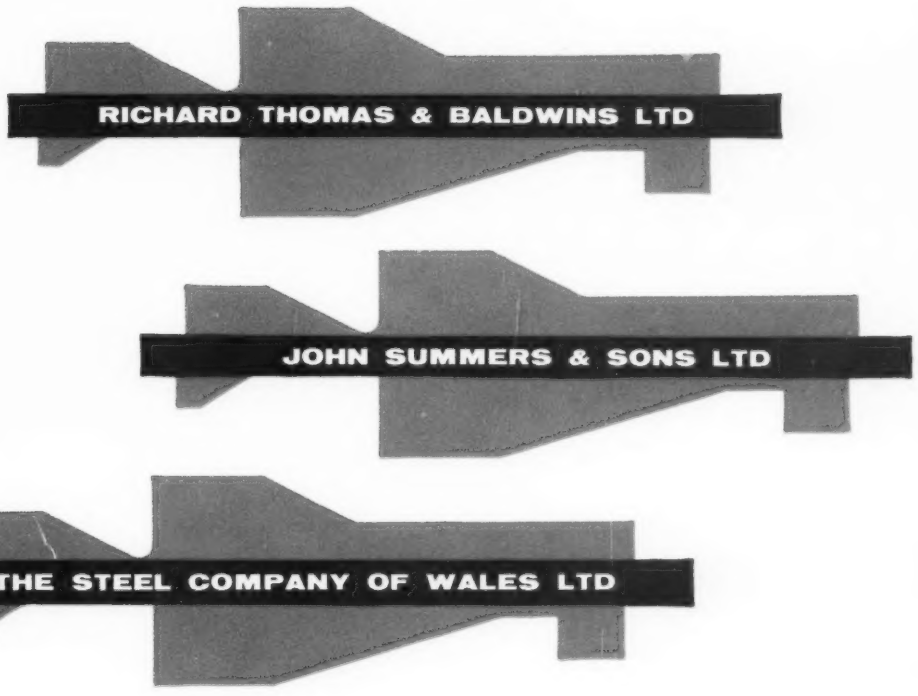
# WIDE STRIP MILLS

## are served by

## "RUST" Continuous Slab Reheating Furnaces

## engineered and installed

by S & A




S & A STEIN & ATKINSON LTD. LONDON

Broadway sa30

# Power behind Scottish

# steel

Ravenscraig, near Motherwell in Scotland, is the site of the latest and most important addition to the iron and steel industry of Scotland to which Colvilles Ltd., have already contributed so much.

Here, from an undeveloped site, the first phase of this expansion, contributing more than a thousand tons of iron daily, was put into operation within three years. In this achievement the G.E.C. assumed responsibility for the whole of the power plant. This included the civil work, cooling towers, boilers, three turbo alternators, two turbine driven blast furnace blowers, two turbine driven pumps, and all associated switchgear, transformers and ancillary equipment to control the gas, steam and electric supplies.

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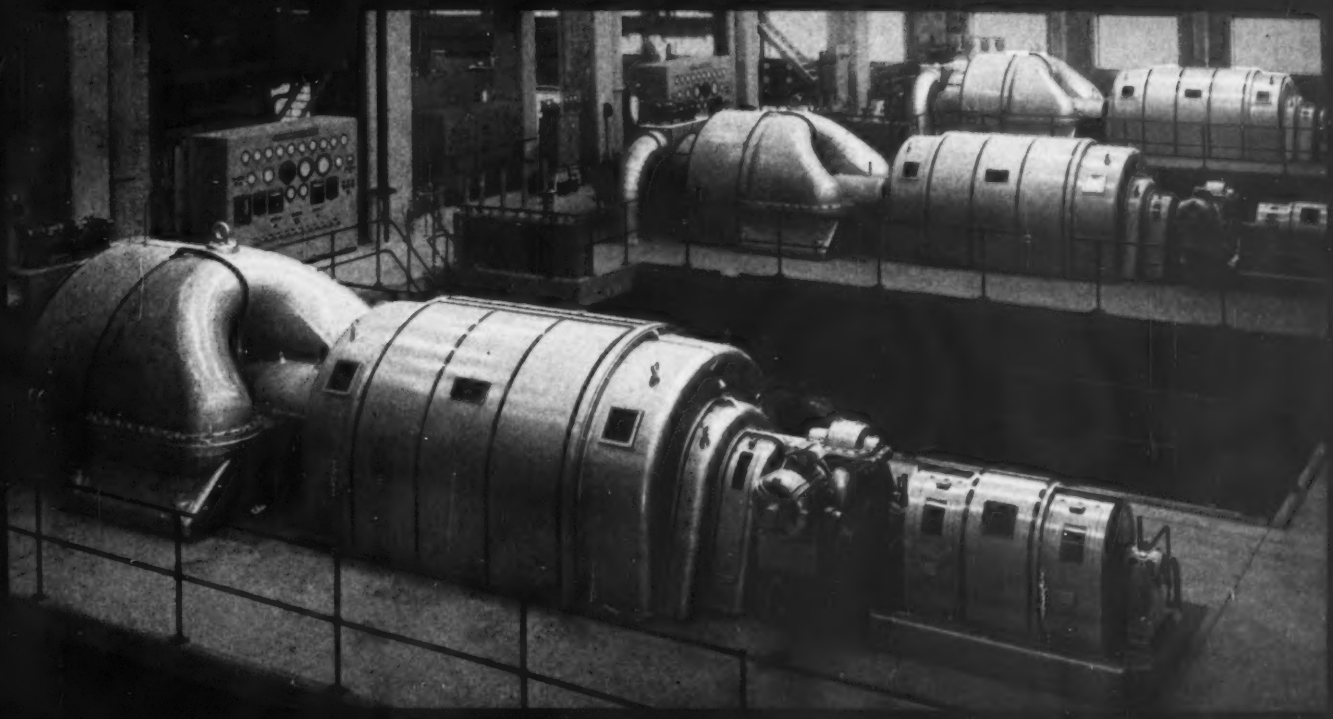

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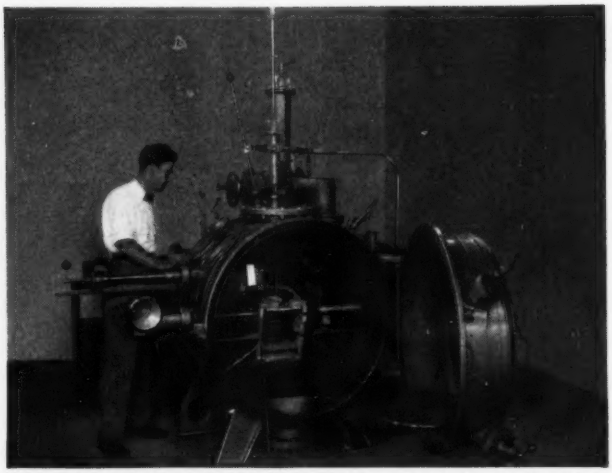

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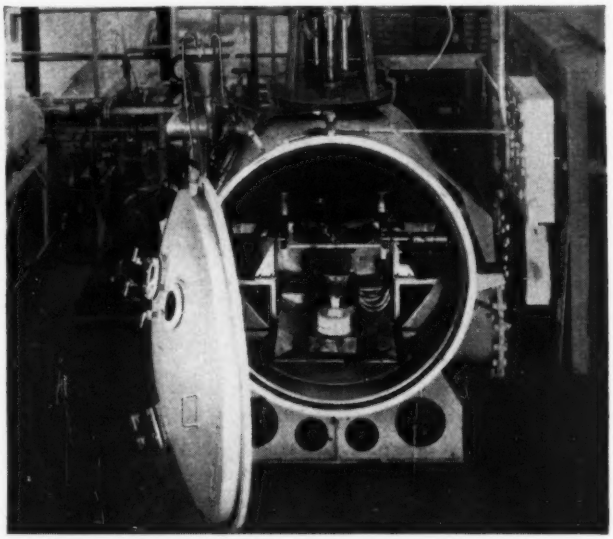

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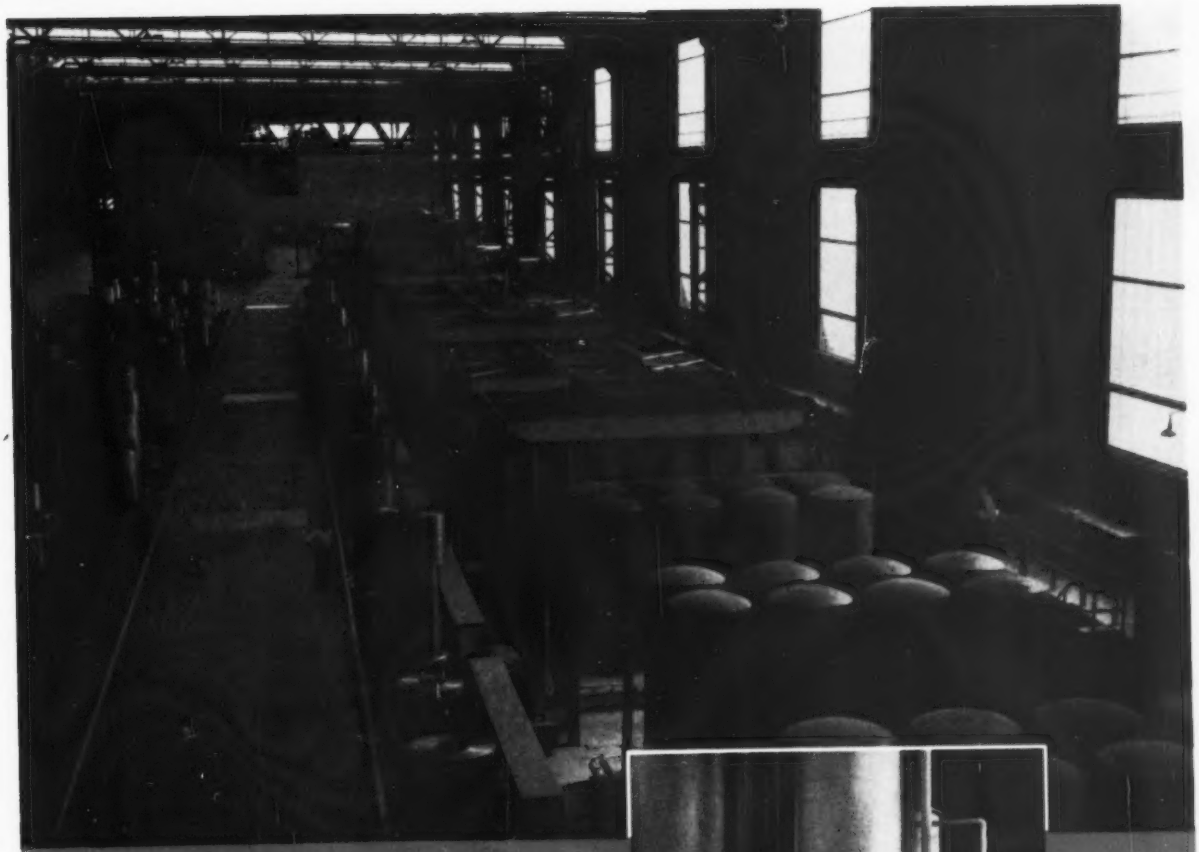
NRC 2



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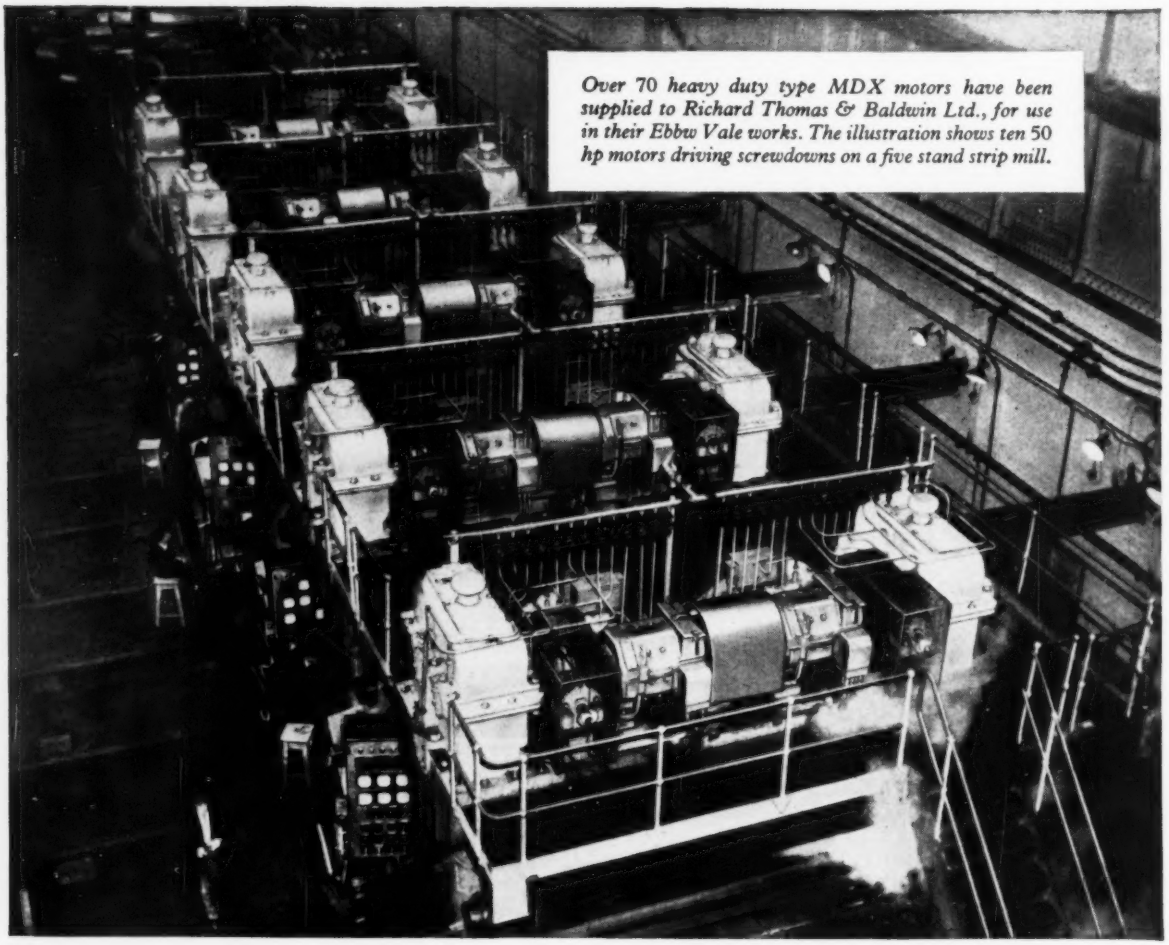

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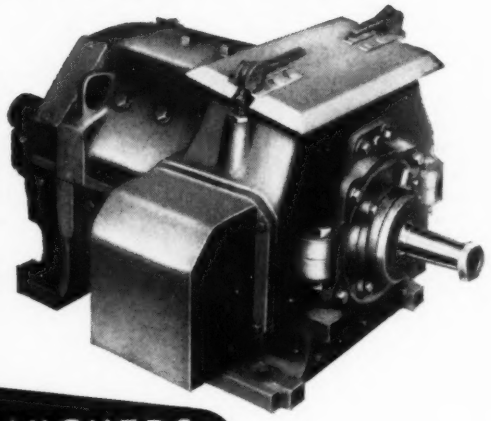

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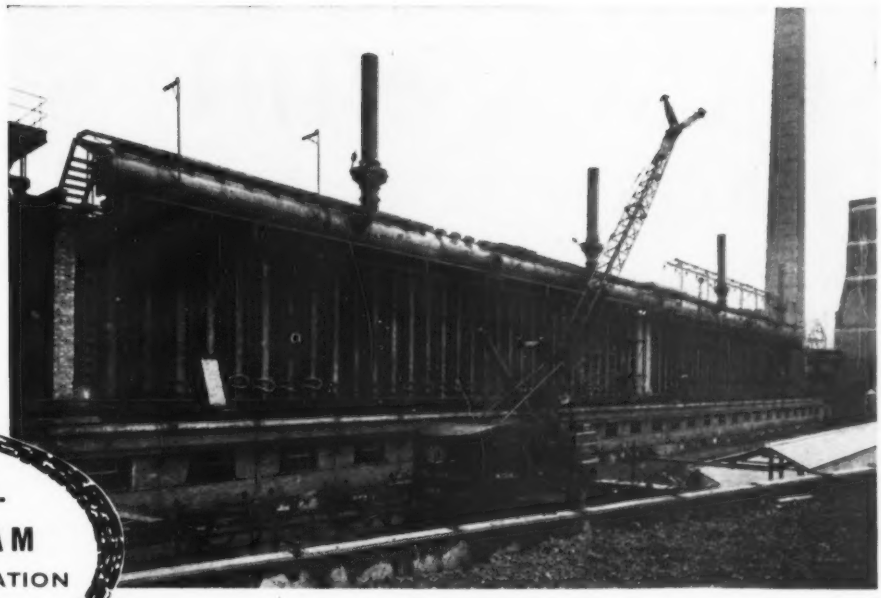
METROPOLITAN-VICKERS

ELECTRICAL CO LTD TRAFFORD PARK MANCHESTER, 17

*An A.E.I. Company*

## Metrovick Motors for Trouble Free Service

J 5702



**APPLEBY-FRODINGHAM**
SECOND INSTALLATION
**66 OVENS**

Operating July 1958

**BAIRDS**
SECOND INSTALLATION
**37 OVENS**

**PRETORIA**
SIXTH INSTALLATION
**51 OVENS**

**MURTON**
FIRST INSTALLATION
**50 OVENS**

**CORBY-DEENE**
FIRST INSTALLATION
**51 OVENS**

**PRETORIA**
FIFTH INSTALLATION
**51 OVENS**

**GLASSHOUGHTON**
FIRST INSTALLATION
**42 OVENS**

**REDBOURN**
THIRD INSTALLATION
**62 OVENS**

**LYSAGHT**
FIFTH INSTALLATION
**23 OVENS**

**VANDERBIJL PARK**
THIRD INSTALLATION
**55 OVENS**

# NEW coking plants under construction

The batteries of the second installation of W-D Koppers Coke Ovens for the Appleby-Frodingham Steel Company now nearing completion.

The carbonisation capacity of the two new batteries of 33 Ovens is 1330 tons of coal per day. The previous installations of W-D Koppers and Koppers Ovens each have two batteries of 33 Ovens wih similar carbonising capacities.

The latest installation includes a new ovens bunker with additions to the coal and coke handling plants and extensions of the by-product, benzole and boiler plants.

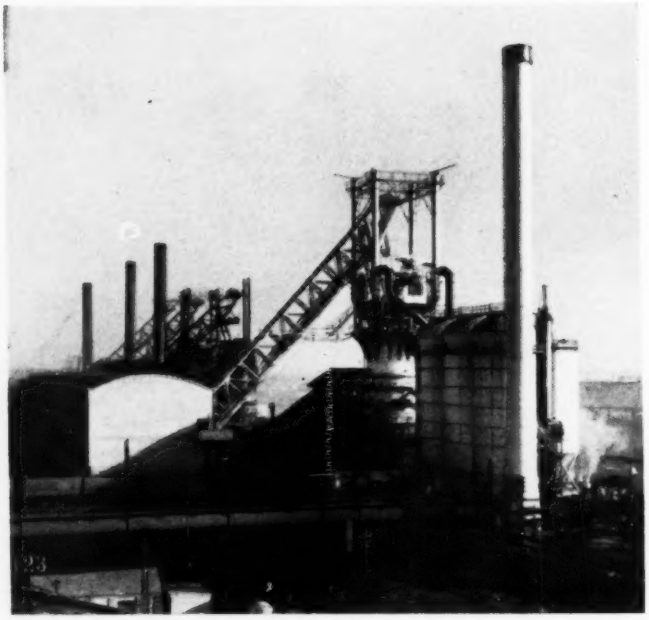
**WOODALL (W-D) DUCKHAM**

***Construction Company Ltd.***

Woodall-Duckham House, 63-77 Brompton Road, London, S.W.3. Tel: KENsington 6355 (14 lines) Grams: Retortical (Southkens) London

# LANCASHIRE STEEL

## MANUFACTURING COMPANY LIMITED



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- SASH AND GLAZING SECTIONS
- HAMMER-LOCK STRUTS
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- IRON, STEEL AND NON-FERROUS CASTINGS
- CONCRETE FLAGS, KERBS, ETC.

LSM

**Manufacturers of Siemens-Martin Open Hearth Basic Steel**

*SPECIALITIES:*

**Wire Rods in all Qualities**

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*for all applications*

Dependability is maintained by manufacturing methods that utilise every mechanical and scientific means to exclude variations in shape and texture.

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in the iron and steel industry
The Eimco 105 Tractor Excavator has proved itself the effective answer for clearing out slag pockets in steel furnaces. Breaking out the slag and brick rubble while still hot, the 105 loads it straight away for disposal without prior blasting in some cases. In addition, the Eimco 105 deals with stockpile loading and general excavating and loading duties.
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you can't beat the tough, heavy duty
EIMCO 105
EIMCO (GREAT BRITAIN) LIMITED
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London Office: PRINCES HOUSE, PICCADILLY, W.1. REGent 2184
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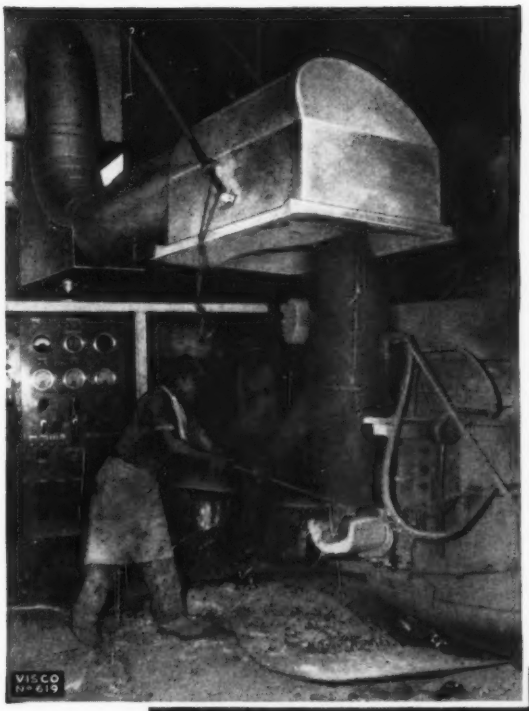
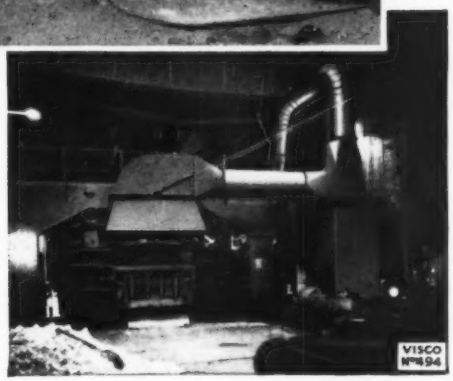



*Famous Firms using "Visco" Arc Furnace Fume Removal Equipment.*

- **Brymbo Steel Works Ltd.**
- **English Steel Corporation Ltd.**
- **Thos. Firth & John Brown Ltd.**
- **Samuel Fox & Co. Ltd.**
- **Rhodesia Copper Refiners Ltd.**
- **Round Oak Steel Works Ltd.**
- **Scaw Metals Ltd.**
- **Richard Thomas & Baldwins Ltd.**
- **Varde Steelworks (Denmark)**

# More Arc Furnace Fume Removal Plants by VISCO

Every minute during the working day well over 800,000 c.f.m. of fume-laden air from Arc Furnaces is extracted by "Visco" Fume Removal Equipment installed by the firms mentioned in the adjoining column. This comprises 21 plants dealing with the fume from 26 Arc Furnaces. By collecting the fume at the point of generation, working conditions in the melting shops are greatly improved. Drivers of overhead travelling cranes do not have their vision impaired by clouds of fume and greater safety is therefore brought about.

One of our engineers would be glad to call by arrangement to discuss fume removal problems of any kind.

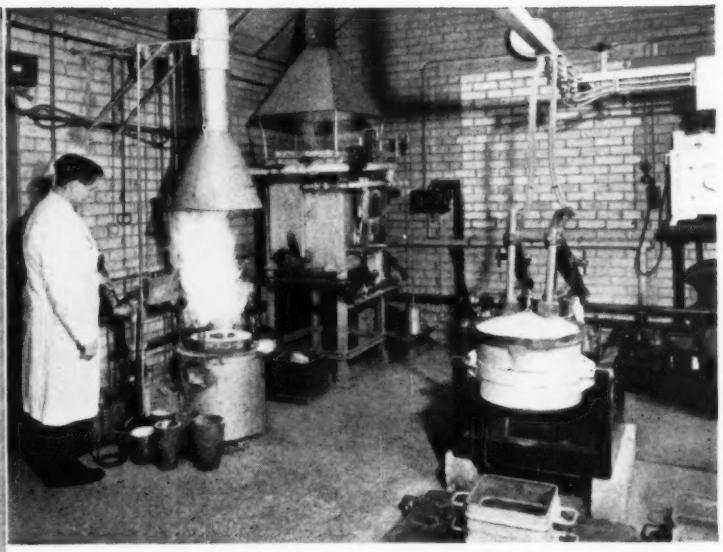
## VISCO
### ARC FURNACE FUME REMOVAL PLANT

**THE VISCO ENGINEERING CO. LTD. STAFFORD ROAD, CROYDON**

*Also makers of water coolers, air filter & dust collectors*

*Phone*: CROYDON 4181

# Finding the answer to your problem



If you should be faced with a technical problem in connection with casting defects in foundry sands, you are invited to call on the facilities of our experimental foundry and sand testing laboratory.
In addition, our foundry technicians will be pleased to visit any foundry man who might wish to discuss a particular difficulty on the spot.
These 'Fulbond' services are offered without charge and are, of course, entirely confidential.
If you will kindly telephone or write to us, we shall be very pleased to co-operate in finding the answer to your problem.

## THE FULLERS' EARTH UNION LTD.

**Patteson Court,**
**Nutfield Road, Redhill, Surrey.**
**Telephone Redhill 3521**

CMF 44

# JMC furnaces . . .

# . . . what you need to know

Platinum wound electrical resistance furnaces have advantages over all other heating methods for temperatures in the range from 1150°C to 1750°C. By progressive improvement in design and construction, Johnson Matthey have over many years developed a comprehensive range of these furnaces and of control units specially designed to suit their characteristics.

Several data sheets describing JMC resistance furnaces in detail are now available, and others will be issued shortly. May we add your name to our mailing list?

*Information regarding the properties of the platinum metals and their industrial applications is freely available.*

A binder to contain the data sheets will be supplied on request.

**Johnson Matthey**

**JOHNSON, MATTHEY & CO., LIMITED**
**73-83 HATTON GARDEN, LONDON, E.C.1**
**Telephone: Holborn 6989**

*Vittoria Street, Birmingham 1.* *Telephone: Central 8004*
*75-79 Eyre Street, Sheffield 1.* *Telephone: 29212*

LA 82

# A BIG BITE

*out of METAL HANDLING COSTS!*

IGRANIC **LIFTING MAGNETS**

***AN IGRANIC MAGNET FOR EVERY LIFTING JOB***

**CIRCULAR LIFTING MAGNETS**

| | |
|---|---|
| 24″ Light duty model | (slabs up to 6,500 lb.) |
| 39″ General service | (slabs up to 12,000 lb.) |
| 45″ Locomotive crane type | (slabs up to 35,000 lb.) |
| 55″ General heavy duty | (slabs up to 47,500 lb.) |
| 65″ Large tonnage model | (slabs up to 60,000 lb.) |
| 77″ Supermagnet for extra high tonnages | (slabs up to 110,000 lb.) |

**RECTANGULAR LIFTING MAGNETS**

| | |
|---|---|
| R1 24″ long<br>R2 40″ long<br>R3 54″ long<br>R4 76″ long | For the efficient handling of regular shapes in iron and steel including sheets, bars, billets, ingots, pipes, etc. |

Also Multi-polar coil-lifting magnets and plate-lifting magnets for fork lift trucks.

The amazing power of IGRANIC magnets means quicker lifts, bigger loads on all ferrous-metal handling jobs from scrap turnings to steel slab, girders to coiled strip. The 77″ supermagnet—biggest in the world—will lift up to 110,000 lb. slabs with ease! Where else can you find such strength—such long life too? For IGRANIC magnets have many exclusive features for extra toughness, trouble-free service. Moisture problems are *out*—for IGRANIC magnets are *waterproof*, with a one-piece monolithic coil and double-sealed terminal boxes. Anti-wear devices prolong lifting chain life too and all parts are ruggedly made to resist rough handling and high temperatures.

***IGRANIC LIFTING MAGNETS***

*First for metal handling for over 40 years*

# IGRANIC

ELECTRIC COMPANY LIMITED

***HEAD OFFICE AND WORKS: BEDFORD ENGLAND***

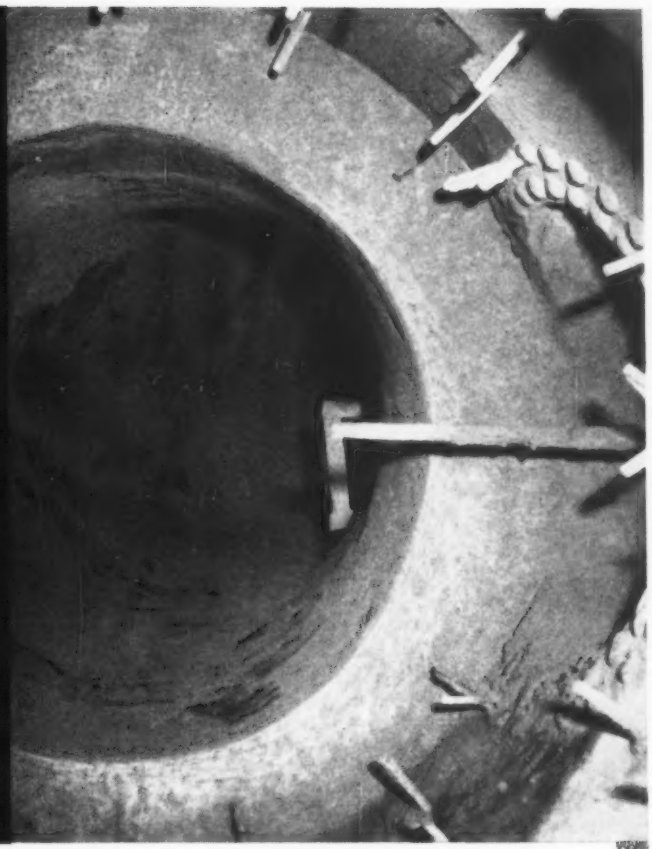
LONDON AND EXPORT OFFICES: VICTORIA STATION HOUSE 191 VICTORIA STREET LONDON SW1

*A Metal Industries Group Company*

DISTRICT OFFICES: BIRMINGHAM BRISTOL CARDIFF EAST ANGLIA GLASGOW LEEDS MANCHESTER NEWCASTLE SHEFFIELD

P.2215/10/95

# Tri-Mor Plastics and Castables make installation Easier Quicker and Cheaper



Bustle main of Queen Victoria Blast Furnace showing 'gunned' lining in Tri-Mor Dense 'Guncrete'. (Inside diameter of shell 5 feet. Thickness of 'Guncrete' lining 9 inches).
*Photograph by courtesy of the Appleby-Frodingham Steel Co. (Branch of the United Steel Companies Ltd.).*

## TRI-MOR GRADES

**TRI-MOR** *Standard Castable*

A medium texture refractory having negligible shrinkage up to 1,350 C. Suitable for casting special shapes or for monolithic structures. Limiting service temperature 1,350 C.

**TRI-MOR** *High Strength Castable*

A similar refractory to Tri-Mor Standard Castable, but specially developed to have very high mechanical strength over the lower temperature range. Maximum service temperature 1,250 C.

**TRI-MOR** *High Temperature Castable*

Suitable for face temperatures up to 1,600 C; has an extremely high resistance to thermal shock; used for cast in situ monolithic structures and for pre-cast refractory shapes; can be applied with a cement gun.

**TRI-MOR** *High Temperature Mouldable*

A plastic refractory for use up to 1,650 C: low shrinkage and a high resistance to spalling. Supplied mixed to the correct consistency for installation.

**TRI-MOR** *Dense "Guncrete"*

A hydraulic setting refractory with a maximum service temperature of 1,300 C. It has a high resistance to abrasion. Designed for application by cement gun, but can be trowelled.

**TRI-MOR** *Insulating Castable*

An insulating castable for maximum service temperatures of 1,200 C; low thermal conductivity is its main feature.

**TRI-MOR** *Insulating "Guncrete"*

Similar to Tri-Mor Insulating Castable but for application by cement gun.

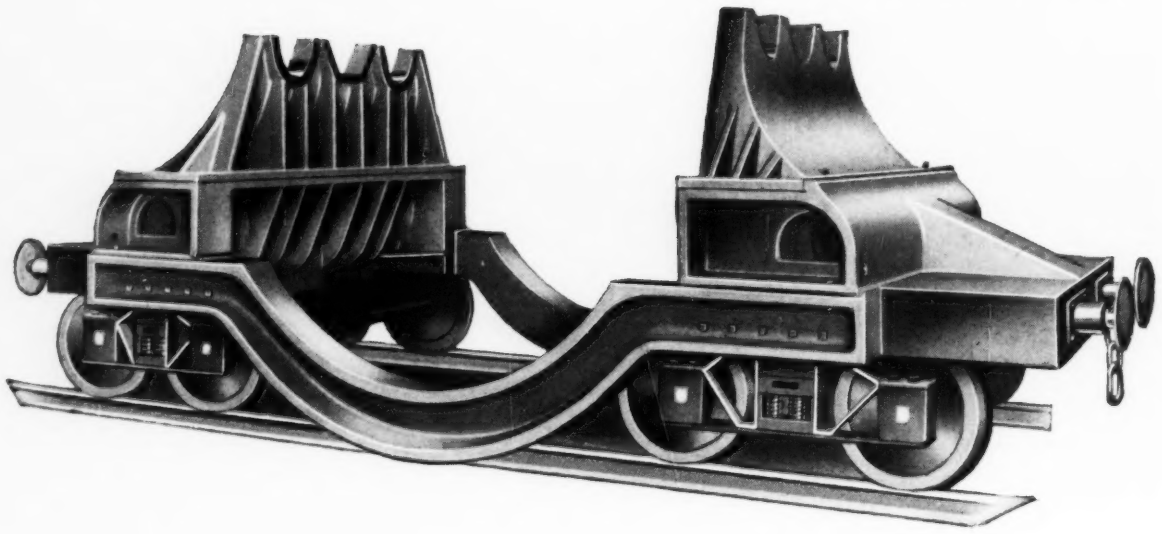
*Full details of each grade are available on request.*

MORGAN Refractories Ltd

*For further information write to:* MORGAN REFRACTORIES LTD NESTON, WIRRAL, CHESHIRE. TEL: NESTON 1406 NE 143

# FABRICATED CARRIAGE BY Thornton

**Engineers and Contractors**



**75-ton Hot Metal Ladle Carriage**

We design and manufacture:

Equipment for Coke Ovens, Blast Furnaces and Open Hearth Plant. Rolling Mills for the Ferrous and Non-Ferrous Industries and Rolling Mill auxiliaries.

Plate work.

Bogies of all kinds for carrying light or heavy loads.

General Engineering work of various kinds.

*We shall be glad to quote against your requirements and specifications.*

**B. THORNTON LIMITED**

Turnbridge, Huddersfield. Phone. Huddersfield 7541

# *Staffordshire Landscapes*

Scenes such as these are common in South Staffordshire. A century lies between the prime years of the foreground, and the present day development behind.

Though the occupants of the new houses may give little thought to the sunken canal boat, the progress which began its decay may also have brought their forefathers to the district to share, and to play their part in the swelling industrial activity which was growing from the pioneering spirit of the early Iron Masters.

Progress in Staffordshire, bringing almost every major improvement in the technique of iron-founding since 1700, is apparent in the landscape of the County as well as in its products.

*For almost a century and a half Pig Iron has been manufactured at Darlaston Iron Works. Today, the most modern methods of metallurgical control of raw materials and the finished product, enables them to supply Pig Iron of consistent uniformity to the most exacting specification.*

**Bradley & Foster LIMITED**

FOR QUALITY CONTROLLED
REFINED PIG IRON

DARLASTON · STAFFORDSHIRE

*A member of the Staveley Coal & Iron Co., Ltd., Group.*

# NEWALLS INSULATING BRICKS

## HOLD THE HEAT PROFITABLY . . .

Today, burning fuel economically has become a national duty—and even better business.

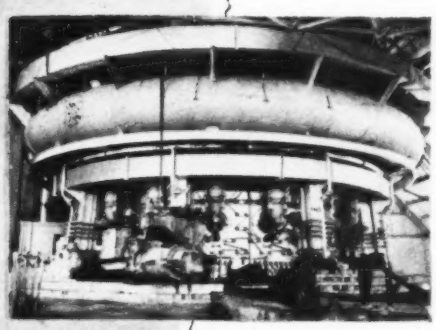
No one can afford to let fuel burn uncontrolled and let heat escape without working. It is more profitable to use Newalls insulating bricks to hold the heat in place, and to use them again and again, until long after they have paid for themselves. There are many obvious advantages. By using every degree of available heat, and maintaining even burning temperatures you cut down fuel consumption and wastage of materials.

By putting your insulation problems into the hands of Newalls you can increase your fuel economy and plant efficiency with lasting effect.

Newalls
(REG'D BRAND)

**NONPAREIL and NEWPAREX INSULATING BRICKS**

NEWALLS INSULATION CO. LTD.
*Head Office:* WASHINGTON, CO. DURHAM.
*A Member of the* TURNER & NEWALL ORGANISATION
*Offices and Depots at* LONDON, GLASGOW, MANCHESTER, NEWCASTLE UPON TYNE, BIRMINGHAM, BELFAST, BRISTOL AND CARDIFF. *Agents and vendors in most markets abroad.*



## iron making plant at Aviles

Head Wrightson were responsible for the complete Iron Making Plant comprising Ore Handling, Blending, Sintering, Blast Furnace and Gas Cleaning Plants, designed to produce over 1,000 tons of pig iron per day. Participation in this iron and steel works project demonstrates Head Wrightson's ability to undertake major contracts of this nature at home and abroad.

# HEAD WRIGHTSON

IRON & STEEL WORKS ENGINEERING CO LTD

THORNABY-ON-TEES · LONDON · JOHANNESBURG · TORONTO · SYDNEY · CALCUTTA

**MARGAM ELECTRO FINISHES LTD**

# MEFCO

**BONDERIZED** **ELECTRO-ZINC COATED**

# STEEL SHEETS

for MAXIMUM PROTECTION

- Bonderized to provide paint adhesion
- Clean to handle
- Rust-resistant surface
- Adds substantially to product life
- Sales and design advantages
- Increases die life
- Protection during fabrication
- Draws, bends and forms readily
- Reduces surface-finishing costs

***RANGE:***

| | |
|---|---|
| Sheet Gauges | 24-10 B.G. |
| Plate up to | 3/16 in. |
| Maximum Sheet Width | 60 in. |
| Maximum Sheet Length | 180 in. |
| Coating Thicknesses | 0.00005 in. to 0.0005 in. Zinc |

ONE-SIDE COATINGS ALSO AVAILABLE
and ELECTRO-ZINC COATED
AND SEALED STEEL SHEETS

*SELLING AGENTS:*

**THE STEEL COMPANY OF WALES LTD**

HOME TRADE ENQUIRIES: ABBEY WORKS, PORT TALBOT, GLAMORGAN

*Telephone: Port Talbot 3161 · Telegrams: Steel, Port Talbot*

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*Telephone: Trafalgar 4300 · Telegrams: Coldrold, Piccy, London*
*Cables: Coldrold, London*

MEFCO *Registered Trade Mark* *MARGAM ELECTRO FINISHES LTD*

*This is one of a series of advertisements by* ***RITCHIE CALDER*** *which is appearing in the National Press to illustrate how mechanical handling may help man's muscle emancipation*

# MILLIONS OF MEN WITH TEASPOONS!

LIKE GNOMES in some weird, exotic fairy-tale, copper-red bodies were swarming up swaying bamboo ladders with baskets of rubble on their heads. Others, in parasol hats, were crouched over great boulders which they were breaking up with chisels and wedges into pieces small enough for the basket carriers. This was in central Java, where the peasants were hewing a canal through the volcanic mountains. Along the canal would eventually flow the waters of the Progo River, to cascade over the terraces of The Hungry Hills. Here rice would grow where it had never been grown before.

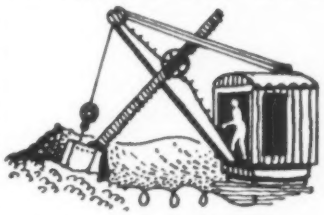
## SAME METHODS AFTER 2,500 YEARS

My brain, teetering in the sweltering heat, tried to grasp that this was the mid-Twentieth Century; this freakish effort of flesh-and-blood and primitive iron was how, as long ago as 542 BC, the slaves must have dug the Hillah Canal in Ancient Babylonia. Two thousand five hundred years later, the Javanese peasants had taken twelve months to cut 2½ miles of their 40 mile canal.

With picks and mattocks, hammers and chisels, they were engaged in a desperate clod-by-clod battle with the earth to dig that canal and get more food. With a few sticks of gelignite, mechanical diggers, hoppers and bulldozers, they could have had it in as many months as the years they were now taking.

It reminded me of an incident in the 1930's when I was standing on a bridge near London watching a giant steam-shovel biting off tons of earth to each 'mouthful', digging a railway cutting. Unemployed men were watching too and one of them said bitterly: "If it weren't for that ruddy machine, there would be work for hundreds of men with picks and shovels". "Or", said another bystander, "millions of men with teaspoons".

But mechanical handling is the winning of time. Modern developments—the deep foundations of tall buildings, the clearing of sites for new towns, the great hydro-electric schemes, could not be contemplated without it. "Millions of men with teaspoons" could never have excavated the vast new Chapel Cross atomic energy station—a station that will produce a unit of atomic power costing one-twentieth of the same unit in muscle power.



***MITCHELL ENGINEERING*** *are designing and constructing Mechanical Handling Plant to meet world needs in the New Atomic Age*

MITCHELL

MITCHELL ENGINEERING LIMITED ONE BEDFORD SQUARE LONDON

# Control Gear for Steelworks Plant

Main Panel and Control Stations for the Main Motor Screw Downs and Coilers of single stand Reversing Mill.
Supplied to The English Electric Co. Ltd., Stafford, for Messrs. Graham Firth Steel Products Ltd., Walsall.

*Contactor Switchgear Limited manufacture automatic electric control gear (up to 6,600v. A.C. and 650v. D.C.) for Power Station Auxiliaries, Steel Works, Water Works, Rolling Mills, Sewage Works, Chemical Plants, Gas Works and Cement Works including Cranes, Electric Furnaces, Machine Tools, Fans and Pumps.*

**May we have *your* enquiries for steel works automatic control gear?**

CONTACTOR SWITCHGEAR LTD
WOLVERHAMPTON ENGLAND

## CONTACTOR SWITCHGEAR LTD.

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Where there are steelmakers...
...there's
ADAMSON-ALLIANCE
STEELWORKS EQUIPMENT
LADLE, SOAKING PIT & STRIPPING CRANES
OVERHEAD TRAVELLING CRANES • CHARGING MACHINES • FORGING MANIPULATORS
ADAMSON-ALLIANCE CO. LIMITED
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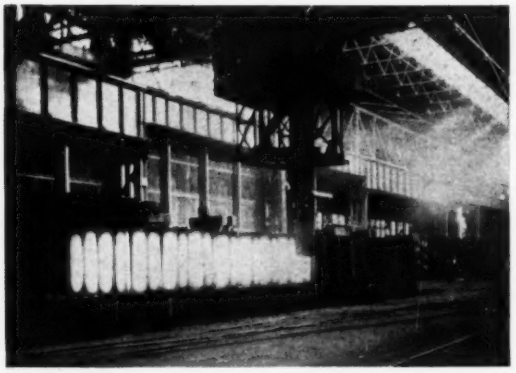
# JOHN LYSAGHT'S SCUNTHORPE WORKS LTD.

## NORMANBY PARK STEELWORKS. Tel.: SCUNTHORPE 2271

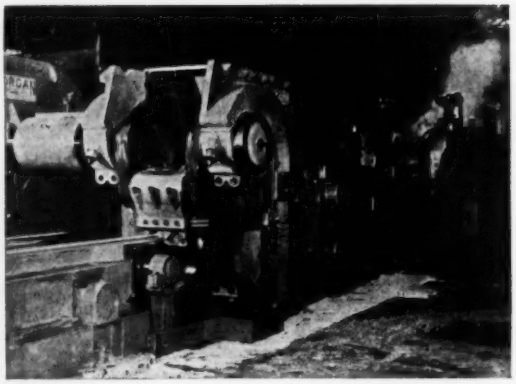
**BASIC OPEN HEARTH STEEL PRODUCERS**

QUALITIES

**CARBON STEELS UP TO ·75**
**DEEP STAMPING AND RIMMING**
**SILICON AND SILICO-MANGANESE**
**FREECUTTING**
**LEAD-BEARING**



STRIPPING BAY



MORGAN MILL FLYING SHEARS

SIZES:

BLOOMS 5″ UP TO 9″

BILLETS 2″, 2¼″, 2½″, 3″, 3½″ & 4″

SLABS 5″ TO 16″ WIDE X 1¾″ TO 2½″ THICK

SHEETBARS 12″ X ¼″ TO ¾″

ROD MILL SIZES

ROUNDS 5G. TO $\frac{7}{16}$″

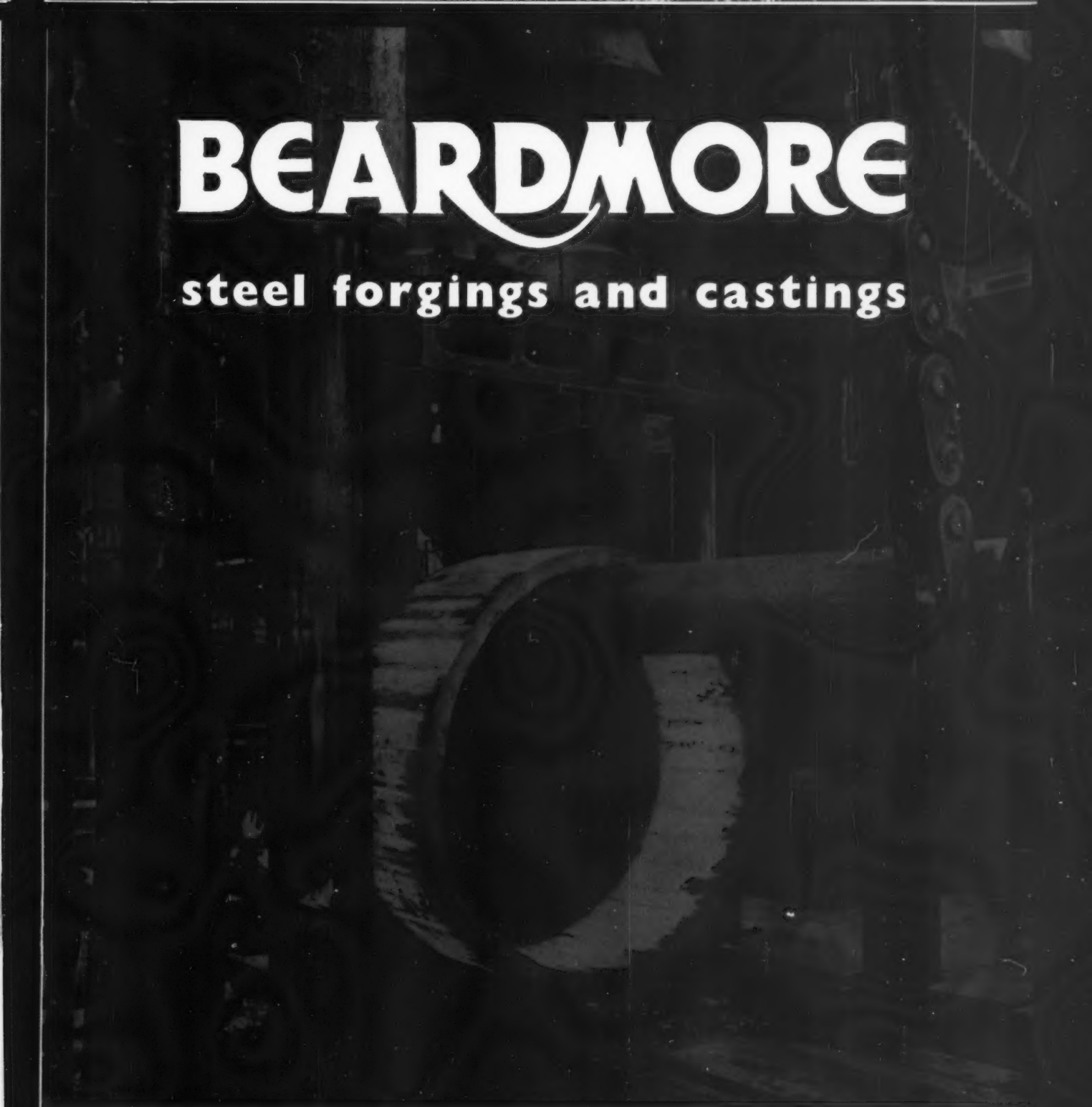
SQUARES 6G. TO $\frac{7}{16}$″

RODS IN 500LB. COILS
(I/D 28″ O/D 36″)

ROD CONVEYOR

# BEARDMORE

## steel forgings and castings

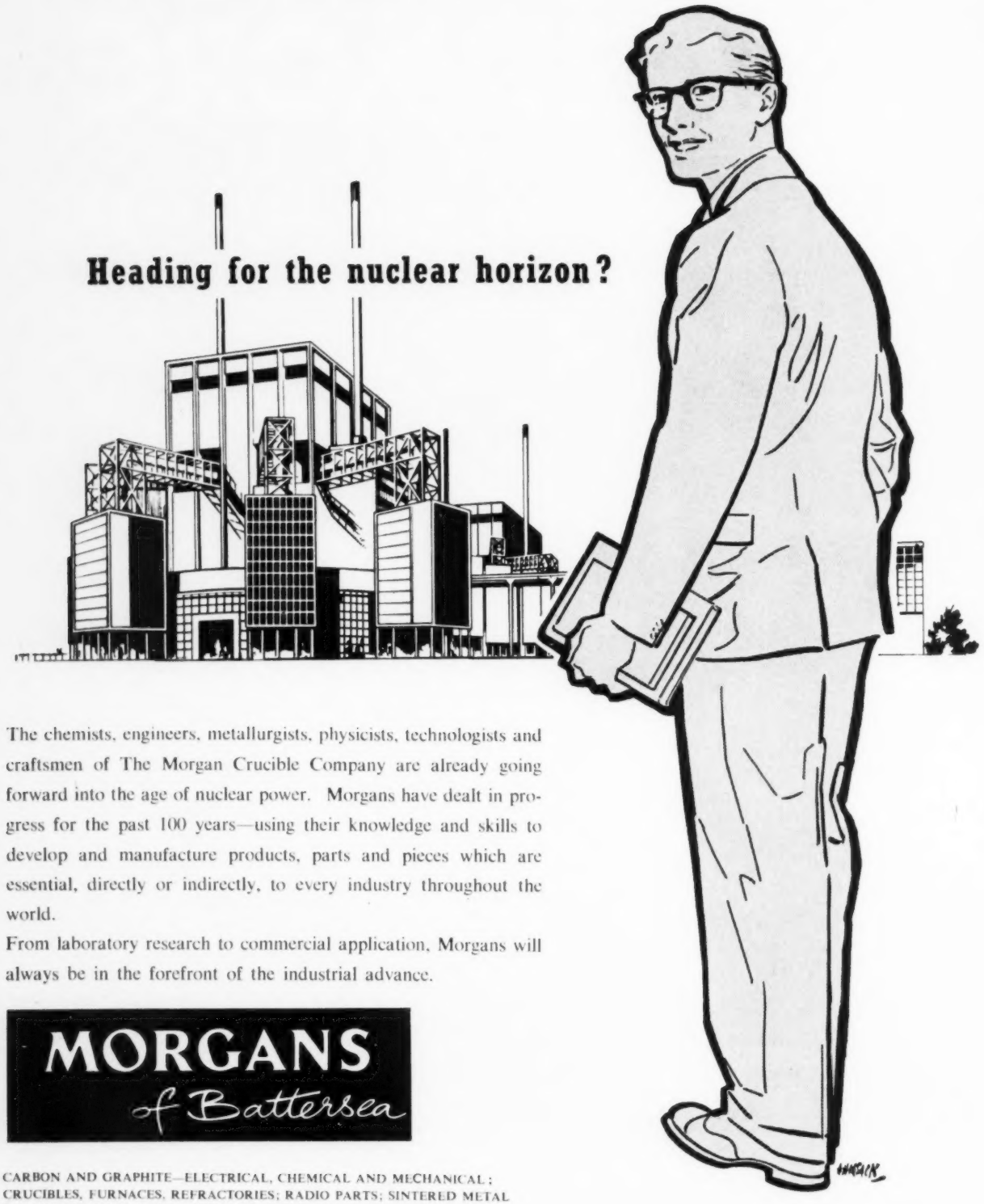


Forgings and Castings in Carbon and Alloy Steels, including Forged Steel Fully Hardened Work Rolls, Forged Steel Solid and Composite Back-up Rolls.

*Illustration:* Gear Wheel Rim Forging.

*Finished Dimensions:* Outside diameter 11′ 0¾″. Inside diameter 10′ 6¾″. Width of forging 3′ 5″.

**WILLIAM BEARDMORE & CO. LIMITED**
Parkhead Steel Works, Glasgow, E.1
*Also at* 11 Hamilton Place,
Piccadilly (Hyde Park Corner), London, W.1
*and* 2 St. John Street, Manchester, 3



## Heading for the nuclear horizon?

The chemists, engineers, metallurgists, physicists, technologists and craftsmen of The Morgan Crucible Company are already going forward into the age of nuclear power. Morgans have dealt in progress for the past 100 years—using their knowledge and skills to develop and manufacture products, parts and pieces which are essential, directly or indirectly, to every industry throughout the world.

From laboratory research to commercial application, Morgans will always be in the forefront of the industrial advance.

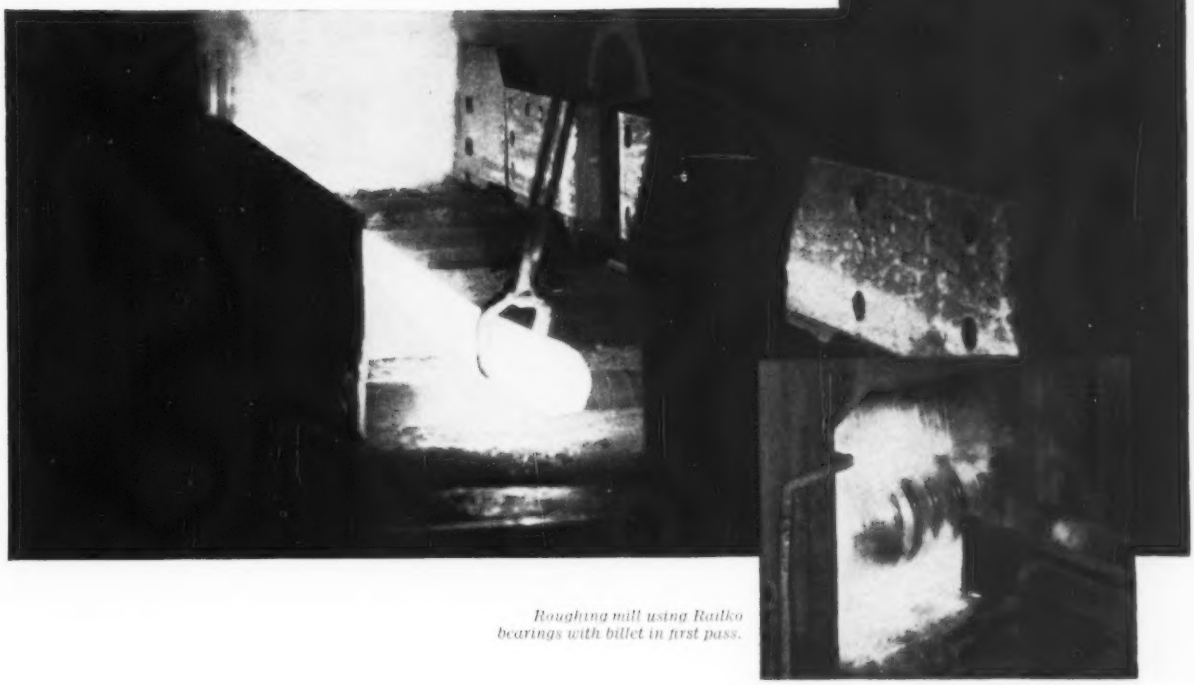
**MORGANS** *of Battersea*

CARBON AND GRAPHITE—ELECTRICAL, CHEMICAL AND MECHANICAL; CRUCIBLES, FURNACES, REFRACTORIES; RADIO PARTS; SINTERED METAL PRODUCTS AND ELECTRIC FURNACE ELEMENTS

THE MORGAN CRUCIBLE COMPANY LIMITED, BATTERSEA CHURCH ROAD, LONDON, S.W.11. BAT. 8822

MCC 124 A

# Railko Bearings at Skinningrove—

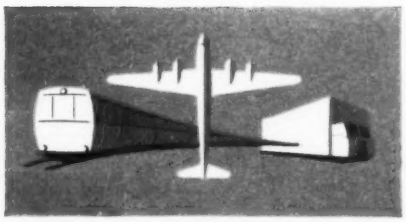


*Roughing mill using Railko bearings with billet in first pass.*

## TOUGH JOB—TOUGH BEARINGS . . .

At Skinningrove, as at many other leading steel works, Railko bearings have proved far more satisfactory where operating conditions are difficult than bearings of conventional materials.

Why do Railko bearings do a better job? . . . they are tough resin-impregnated fabric laminates specially moulded under heat and pressure to give specific properties depending on the work involved. Railko bearings have many advantages. They withstand extreme pressure and movement * have a high strength-to-weight ratio a high compressive strength * a low co-efficient of friction * can be lubricated with a variety of fluids * in certain applications can be used without fluids are resistant to acids and alkalis * have a low rate of wear * are easily machined have excellent weathering properties.



Railko bearings are now widely used by manufacturers in the steel, railway, aviation, engineering and chemical industries. Perhaps they can help *you*? Write to us and we shall be glad to send you full technical details.

Railko bearings help every industry-**Land - Sea - Air**

**RAILKO LIMITED** Railko

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A member of the **Birfield Group**

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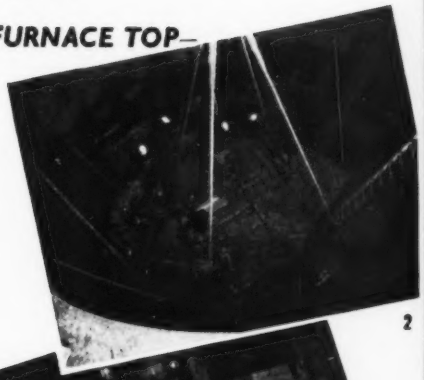
FITNESS FOR PURPOSE STEELS

COLVILLES LIMITED 195 WEST GEORGE STREET · GLASGOW C 2

# For SPEED and ECONOMY in FURNACE WRECKING use the Firm with experience GRANT, LYON

**FURNACE TOP—**



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3

4

**—TO FURNACE BOTTOM**

1

Scenes of operations at No. 1 Blast Furnace
*(photos by permission of John Summers & Sons Ltd., Shotton)*

| | |
|---|---|
| **Furnace blown out** | 23rd April 1958 |
| **Demolition commenced** | 26th April 1958 |
| **Demolition completed** | 13th May 1958 |

**TOTAL TIME** wrecking burden, lining, hearth and bear **18 DAYS**
*(contract carried out in association with Salem Engineering Co. Ltd.)*

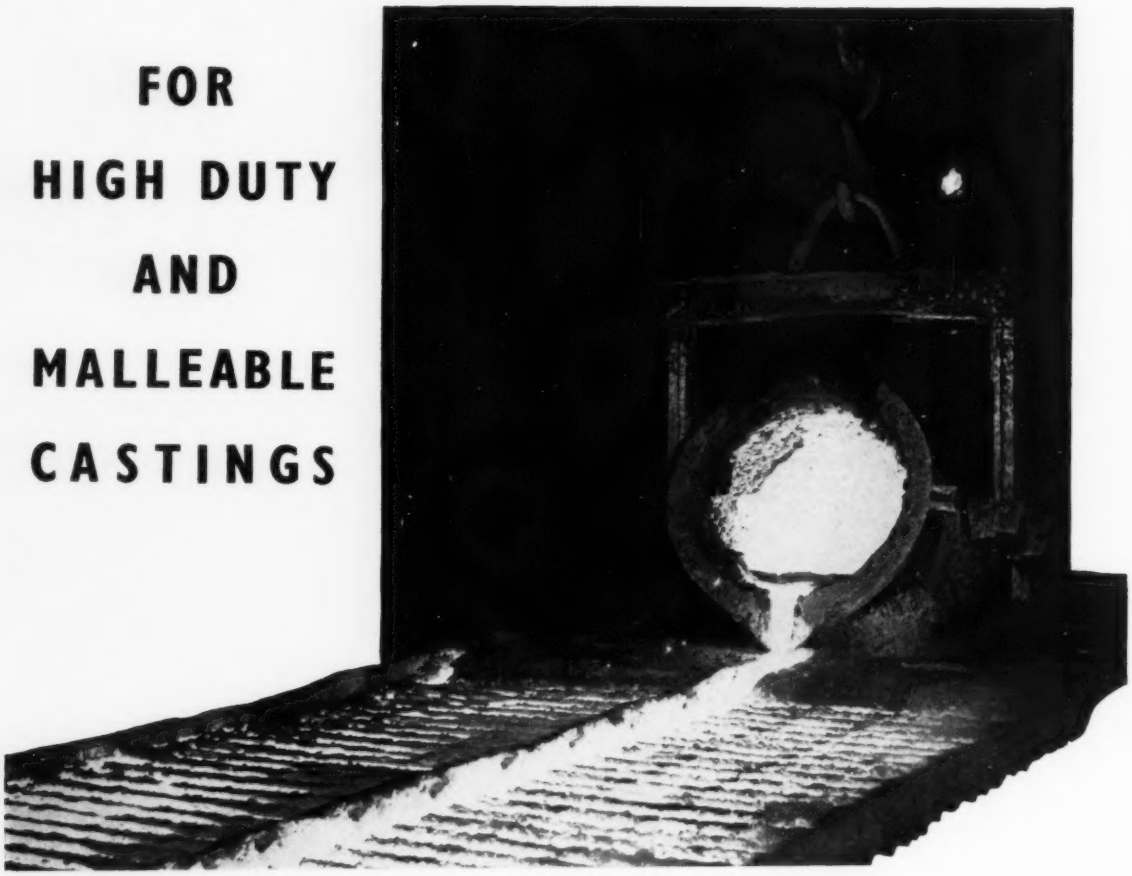
1 *'Lumpy going'—half-way down hearth.*
2 *Men removing throat armour.*
3 *Cat Traxcavator commencing digging, Cast House Floor Level.*
4 *The two 'cats' meet—final 'break through'.*

Your enquiries are invited.

**Grant, Lyon**

AND COMPANY LIMITED
RAILWAY AND CIVIL ENGINEERING CONTRACTORS
Head Office : Scotter Road, Scunthorpe, Lincs.
*Phones : Scunthorpe 230819 London Croydon 4605 Manchester Irlam 2837 Cardiff 21683 Airdrie Wayside 277*

# FOR HIGH DUTY AND MALLEABLE CASTINGS



## The seven standard grades of Stanton Dale Refined Pig Iron

| GRADES: | A | B | C | D | E | F | G |
|---|---|---|---|---|---|---|---|
| SILICON % | 1·9–2·1 | 1·4–1·6 | 0·9–1·1 | 2·25–2·5 | 0·9–1·1 | 1·4–1·6 | 1·9–2·1 |
| SULPHUR % | 0·06 | 0·07 | 0·08 | 0·05 | 0·08 | 0·07 | 0·06 |
| PHOSPHORUS % | 0·6–0·7 | 0·6–0·7 | 0·6–0·7 | 0·4–0·5 | 0·25 | 0·25 | 0·25 |
| MANGANESE % | 0·8–1·5 | 0·8–1·5 | 0·8–1·5 | 0·8–1·5 | 0·8–1·5 | 0·8–1·5 | 0·8–1·5 |
| TOTAL CARBON % | 2·7–3·0 | | | | | 2·8–3·1 | |

In addition to the above other grades of Dale Refined Iron can be supplied to meet more exacting demands which call for alloy additions and special processing.

STANTON *Dale Refined* PIG IRON

THE STANTON IRONWORKS COMPANY LIMITED NEAR NOTTINGHAM

# When you reflect upon MODERN COMFORTS—

## think of R·T·B

But for our pioneering of the continuous-strip mill in this country, there would not be enough steel sheets to meet the present scale of car production.

Our contribution to modern comforts does not stop there, for our products are to be found in every kind of modern household equipment—in streamlined cookers and refrigerators, in easily-cleaned stainless steel, in radio sets, electric cleaners—even in electric radiators.

In fact, there is hardly anything you do—anywhere you go—without enjoying the unseen help of RTB.

*The decoration at the top of the page is a reminder that RTB activities, so deeply associated with sheet steel and tinplate, also include the production of the alloy steels used for vehicle springs, crankshafts and other vital parts.*

## Richard Thomas & Baldwins Ltd.

# HOT ROLLED STEEL BARS

*Round • Square • Flat*

**Ferro-Concrete Bars** *bent to Specification*

# HOT ROLLED STEEL HOOPS AND STRIP

**Coils or cut lengths**

# COLD ROLLED STEEL STRIP

*in all qualities including*

**Special Deep Stamping**
**in cut lengths or coils**
**Electro-Galvanised Strip**

*Speciality:*

**VERY HEAVY CONTINUOUS LENGTH COILS**

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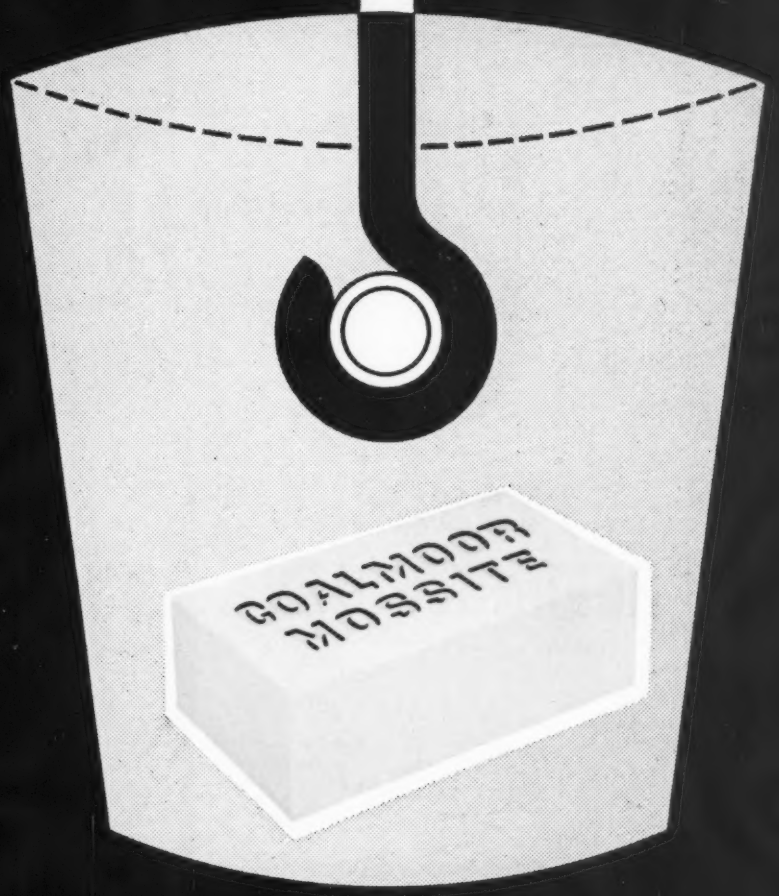
STEIN

*in the lead...*




**COALMOOR MOSSITE**

COALMOOR REFRACTORIES ( HORSEHAY ) LTD. Lightmoor, Dawley, Shropshire

# STEIN *Refractories*

## CEMENTS, PLASTICS AND CASTABLES

From our extensive range of Refractory Cements, Plastics and Castables, covering a wide range of properties, we can supply the correct material for most industrial applications.

### REFRACTORY CEMENTS

***"Nettle" Refractory Cement***
Heat-setting

***"Stein" High Alumina Cement***
Heat-setting

***"Maksiccar 11" Refractory Cement***
Air-setting

***"Maksiccar" Fire Cement***
Air-setting

### PLASTIC REFRACTORIES

***"Maksiccar Patch"***
Air-setting

***"Stein 73 Patch"***
Air-setting

### CASTABLE REFRACTORIES

***"Stein Refractory Concrete"***
Hydraulic-setting

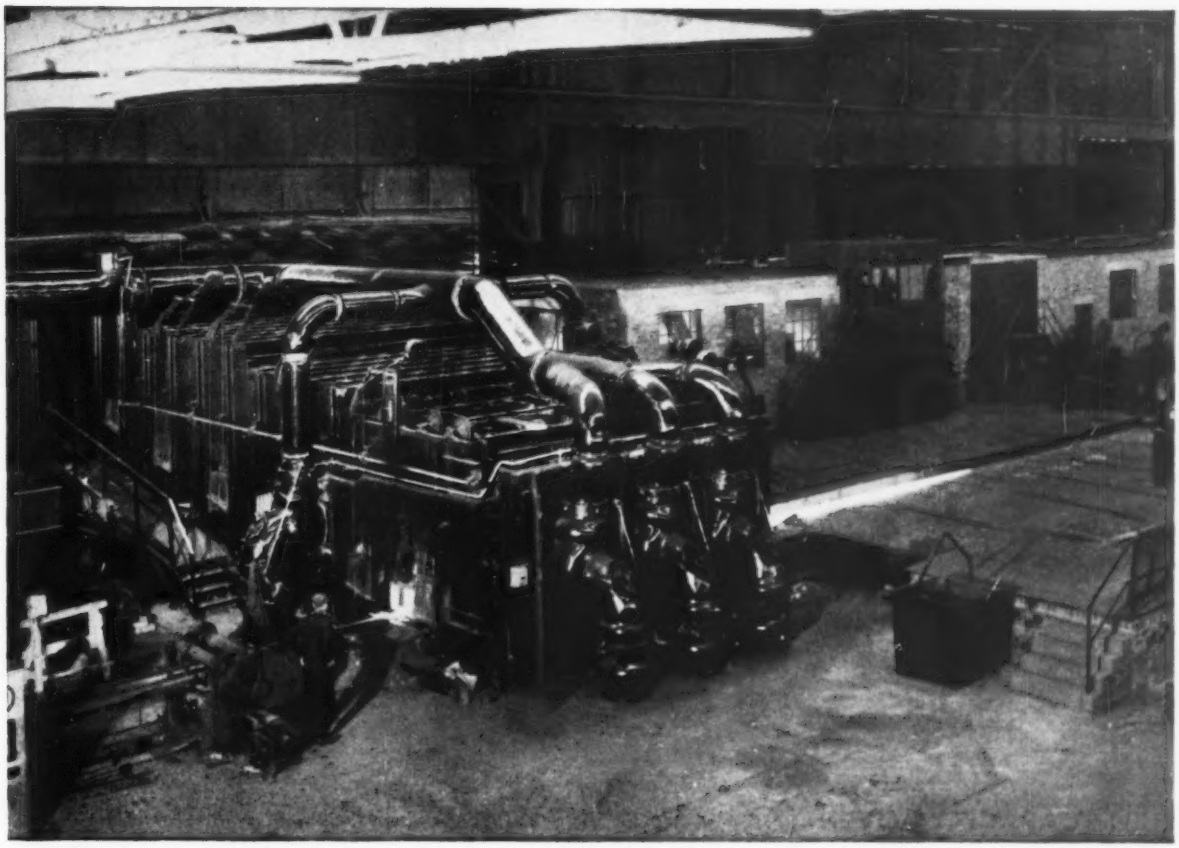
***"Stein Super Refractory Concrete"***
***Nos. 16, 17 & 18.*** Hydraulic-setting

***"Stein Chrome Concrete"***
Hydraulic-setting

**Please ask for a copy of our No. 4 Pamphlet.**

**JOHN G. STEIN & CO. LTD. Bonnybridge, Scotland**

TEL: BANKNOCK 255 (4 LINES)



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Equally suitable for crude producer gas, blast furnace, coke oven, mixed or town gas

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## Super Grade Low Alumina Silica Brick

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FINE SILICA CEMENT FOR SETTING SILICA BRICK

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STIGNIC CEMENT FOR BASIC STEEL LADLES

KILN CAPACITY OVER 1½ MILLION BRICKS

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SPECIAL FIRE CEMENTS for all purposes

•

STEPHENS' SPECIAL ELECTRIC FURNACE ROOF BRICKS

REGENN BRICK without doubt the best Brick for Soaking Pits, Checkers, and Regenerator Chamber Walls

•

**Stephens** SILICA BRICK CO., LTD. KIDWELLY

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Liebers & Marconi

*For steel in any shape or form you'll find ...*

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*A handy GROUP to know*

**GKN products and services include:** INDUSTRIAL STEEL; RAILWAY, COLLIERY AND MINING EQUIPMENT; WHEELS AND CHASSIS FRAMES; SCAFFOLDING AND RAINWATER GOODS; WELDED PIPE-WORK; HOLLOWARE; FORGINGS AND PRESSINGS; COMPLETE ASSEMBLIES OF ALL KINDS; FASTENINGS FOR EVERY PURPOSE.

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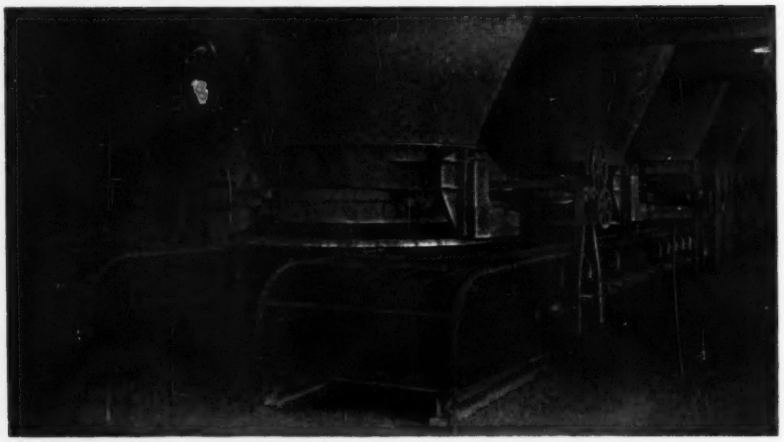

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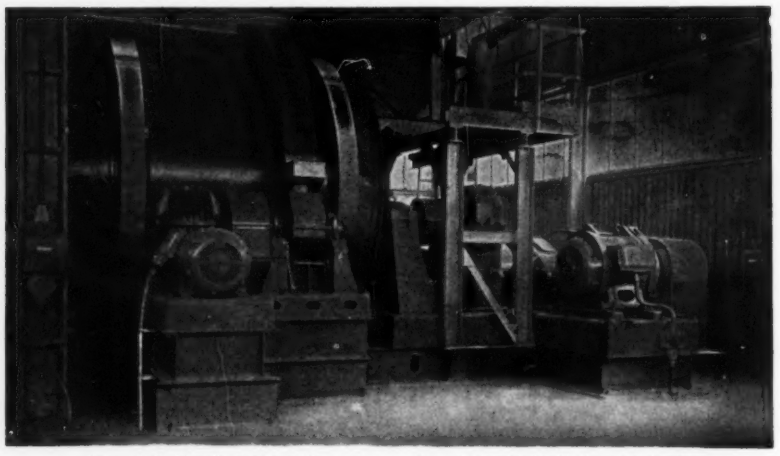

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# Journal of The Iron and Steel Institute

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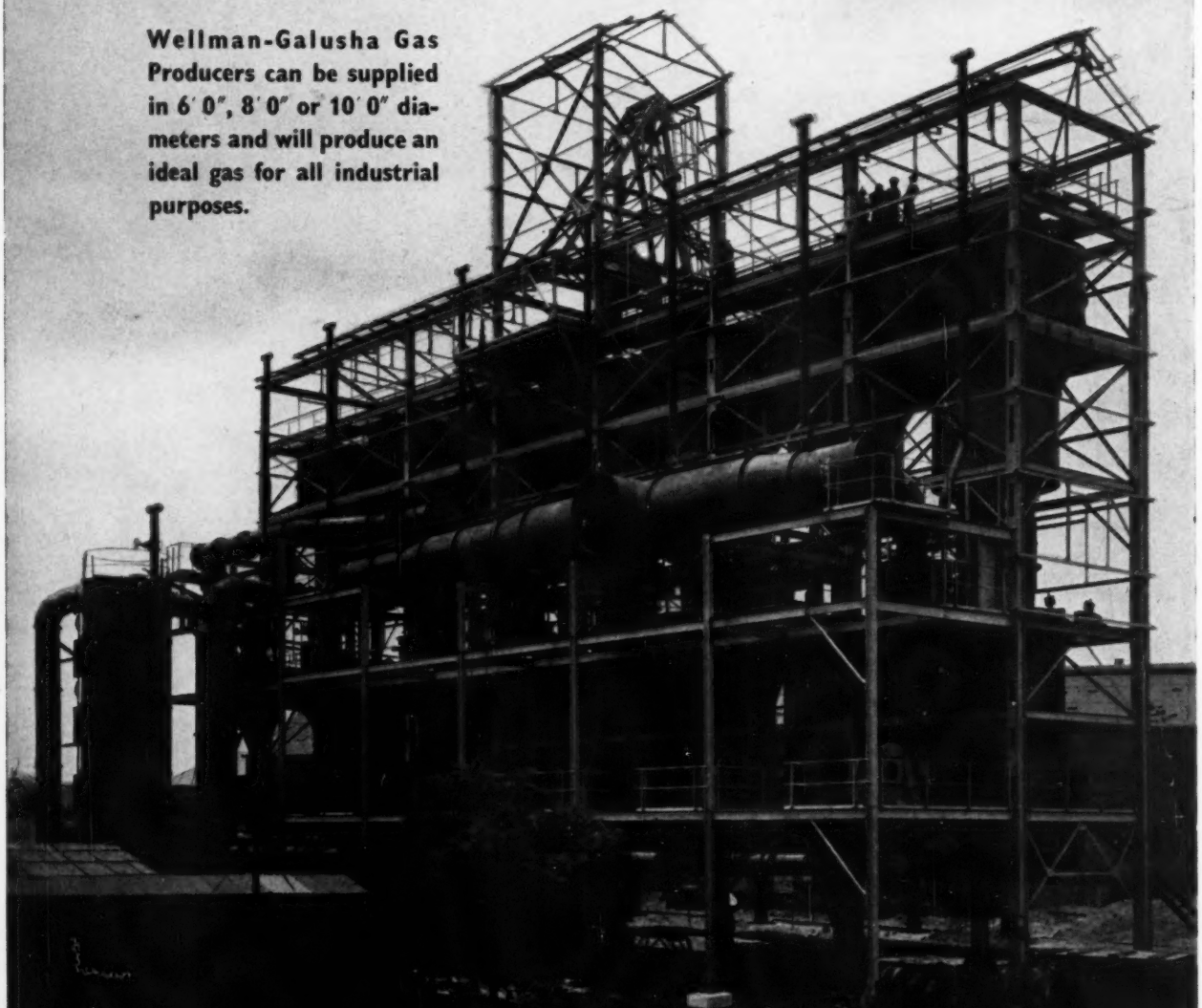

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## N. P. ALLEN, M.Met., D.Sc., F.I.M., F.R.S.

NORMAN PERCY ALLEN graduated in metallurgy at Sheffield University in 1923, and later obtained his M.Met. degree. His work on gases in metals, particularly copper and silver, was started at University College, Swansea, for the British Non-Ferrous Metals Research Association. This work was continued at the University of Birmingham, where he was a member of the staff of the Metallurgy Department from 1928 to 1935, latterly as senior lecturer. He obtained his D.Sc. degree for this work in 1934.

Dr Allen then joined the research staff of The Mond Nickel Co. Ltd. at their Birmingham laboratory, where, amongst other things, he led research teams working on the transformation characteristics, hardenability, and mechanical properties of nickel chromium alloys. The work on low-alloy steels elucidated their transformation behaviour and enabled the rates of transformation to be correlated with hardenability. Investigation of the properties of alloys at elevated temperatures led to the development of the earlier Nimonic creep-resistant alloys.

Since 1944 Dr Allen has been superintendent of the Metallurgy Division of the National Physical Laboratory, first with the rank of Deputy Chief Scientific Officer, and recently with that of Chief Scientific Officer. His interests at the N P L have ranged over a wide field, and have covered such aspects as the deformation and fracture of high-purity iron and iron alloys, the influence of purity in relation to alloy conservation and mechanical properties of heat-treated steels, the development of improved steels for steam power plant, the creep and fatigue of metals, the metallurgy of the newer metals such as titanium and niobium, and the application of new techniques, including radioactive tracers and electron microscopy, to metallurgical problems.

Some of the work on the deformation and fracture of iron, which has been carried out at the N P L over a period of years, forms part of the subject of the eleventh Hatfield Memorial Lecture which he delivered on 9th December, 1958.

Dr Allen has been a member of numerous research committees, and is at present Chairman of the J/E Committee of the British Electrical and Allied Industries Research Association dealing with steels for steam turbines. He has been a member of The Iron and Steel Institute and The Institute of Metals for many years; he is a Fellow of The Institution of Metallurgists, and is currently a Vice-President. He represents The Iron and Steel Institute on the Council of the British Iron and Steel Research Association and is a member of the Research Board of the British Non-Ferrous Metals Research Association. He was elected a Fellow of the Royal Society in 1956.

**N. P. Allen, M.Met., D.Sc., F.I.M., F.R.S.**

*Eleventh Hatfield Memorial Lecturer*

*ELEVENTH*

*HATFIELD MEMORIAL LECTURE*

# The mechanical properties of the ferrite crystal

*By N. P. Allen, M.Met., D.Sc., F.I.M., F.R.S.*

I must first express my deep appreciation of the honour you have done me in asking me to give your Eleventh Hatfield Memorial Lecture. Dr Hatfield was a many-sided personality, and not the least important of his activities was the consistent support he gave to fundamental research in metallurgical topics. His contributions to discussions of the origins of the mechanical properties of metals, and particularly the occasions on which he crossed swords with Dr Rosenhain, are well remembered. In describing the properties of iron as they are now thought to be, I feel that I am entering upon a debate in which he would have been delighted to take part.

My choice of subject was influenced by the thought that there are certain materials whose importance in human economy is so great that they deserve special study, and of these the crystal of alpha-iron is certainly one, for upon its properties the whole of our modern mechanical civilization depends. It had been my first thought to collect and summarize what is known of the properties of this crystal, but I soon found that this was a task more suited to a book than to a lecture, and I had to limit myself first to the mechanical properties of the crystal, and then to those aspects of its mechanical properties that had engaged my personal interest.

## The elastic properties

So far as the theoretical physicist is able to guide us, we are advised that iron is to be regarded as a 'close' metal, in that the cohesion between the atoms is not due to the reactions between the metallic ions and the valency electrons only, but the outer electron shells of the metallic ion come into close contact. Interactions between the 3*d* electrons of the adjacent ions make a considerable contribution to the strength.

In the body-centred cubic structure, this contact occurs in the direction of the cube diagonal (Fig. 1*a*). Regarded from this direction, the structure is seen to consist of rows of atoms in contact, arranged in a slightly overlapping hexagonal pattern (Fig. 1*b*), in which three important types of plane can be distinguished. These planes are respectively {110}, {112}, and {123} planes. Their positions in relation to the unit cube are illustrated in Fig. 1*c*, and the arrangement of atoms in the planes in Fig. 1*d–f*.

One result of this arrangement is that the crystal is very resistant to compression and extension in the direction of the rows of atoms, but less resistant to shear, for the rows of atoms can readily slide over each other. The elastic properties of the crystal are markedly anisotropic. Young's modulus is greatest in the direction of the cube diagonals ⟨111⟩ and least in the direction of the cube edges ⟨100⟩, whilst the shear modulus is least in the direction of the cube diagonals and greatest in the direction of the cube edges.[1]

The structure contains much unoccupied space, and consequently possesses to a marked degree the ability to dissolve small atoms in interstitial positions. The

Delivered at Church House, Westminster, London, on 9th December, 1958.

Dr. Allen is Superintendent of the Metallurgy Division, National Physical Laboratory.

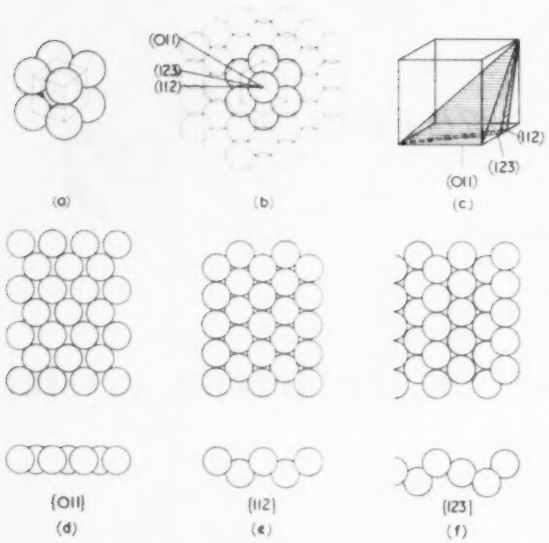



**Fig. 1—The body-centred cubic structure: (*a*) unit cube; (*b*) view along a ⟨111⟩ direction; (*c*) position of (110), (112), and (123) planes having a common [111] direction; (*d*), (*e*), (*f*) arrangement of atoms in {110}, {112}, and {123} planes**

most important open space lies in the position marked *Y* in Fig. 2a, the centre of the edge of the unit cube. Placing a foreign atom in this position involves straining the atoms apart, along the weak direction of the crystal. It is not surprising therefore that carbon and nitrogen when dissolved in iron occupy this position.

If the crystal is unstressed, the interstitial atoms choose any of the positions *Y* without discrimination, but if a tensile stress is applied in a ⟨100⟩ direction, a proportion of the *Y* interstices is widened, and the interstitial atoms tend to move into these wider spaces. If the stress alternates between tension and compression, the dissolved atoms are made to jump in and out of the preferred interstices, and this movement is an important cause of the damping of elastic vibrations.[2]

A particularly marked damping occurs when the relation

$$\frac{\omega}{D} = \frac{3}{2\beta a^2}$$

exists between the frequency of the vibrations and the diffusion constant of the solute, where $\omega$ = frequency of vibration, $D$ = diffusion constant of the solute in ferrite, $a$ = lattice constant, and $\beta$ = numerical constant of order unity, and since the diffusion constant increases with rise of temperature, there is a maximum in the curve of damping capacity against temperature. When carbon is the solute and the frequency is 2 c/s, this maximum occurs at about 50° C (Fig. 2*b*).[3a] By studying the variation of the temperature of this maximum with frequency, it is possible to measure accurately the variations with temperature of the rate of diffusion of carbon in ferrite. Moreover, the magnitude of the internal friction at the maximum is directly proportional to the amount of dissolved carbon in the ferrite and, by studying the variations of this magnitude with heat-treatment, very precise information has been gained on the solubility of carbon in ferrite and the rate at which the carbides are precipitated from a supersaturated solid solution.

Figures 3*a–c* summarize the information that has been obtained in this way.[3] Although 0·018%C is soluble in ferrite at 720° C, the solubility is reduced to 0·0006% at 300° C, and would be expected to be about 0·0000002% at 20° C. The precipitation is so rapid at 300° C that a noticeable proportion of the dissolved carbon is precipitated if during cooling the sample spends as little as 1 s in the neighbourhood of 300° C. At room temperature the precipitation is slower, but begins within a day and is subsequently complete in a few months.

These conclusions are confirmed by microscopic observations. In a slowly cooled high-purity iron containing as little as 0·0025%C, small distinct crystals of cementite can be found (Fig. 4).[4a] When carbon is retained in solution by quenching and then precipitated in a short time at 200–250° C, thin platelets lying in {100} planes are observed (Fig. 5).[5] Precipitation at room temperature is not so easy

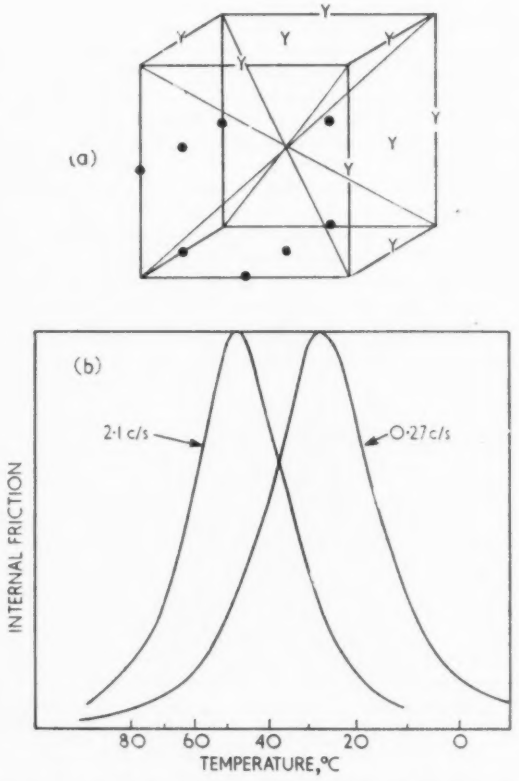



**Fig. 2—Origin of damping by interstitial atoms: (*a*) interstitial position; (*b*) the influence of temperature and frequency on the damping due to dissolved carbon**

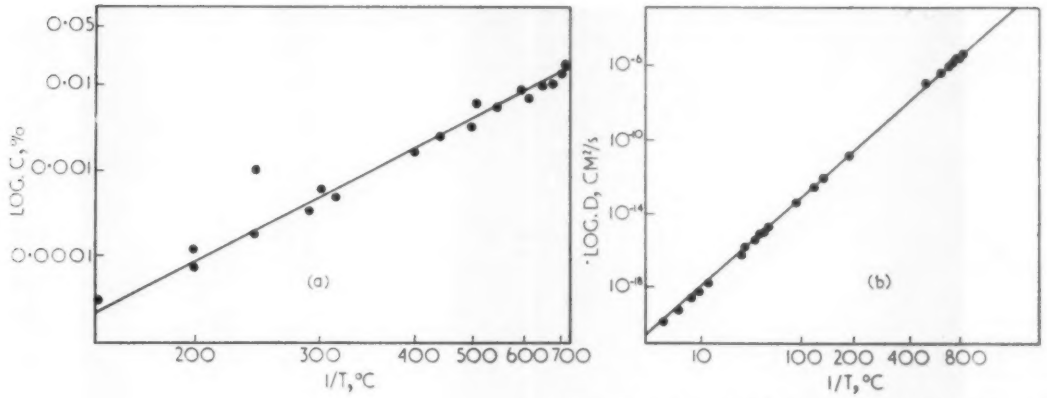



Fig. 3—The behaviour of dissolved carbon: (*a*) variation of solubility with temperature; (*b*) variation of diffusion constant with temperature; (*c*) rate of precipitation of carbides from ferrite

(*c*) Fraction of available carbon precipitated in stated time at stated temperature

| Time | 20° C | 100° C | 200° C | 300° C |
|---|---|---|---|---|
| 1 s | ... | .... | .... | 0·0055 |
| 1 min | ... | ... | 0·028 | 0·76 |
| 1 h | ... | 0·07 | 0·9999 | >0·9999 |
| 1 day | 0·0025 | 0·9993 | .... | ... |
| 1 month | 0·68 | >0·9999 | .... | ... |
| 1 year | >0·9999 | .... | ... | ... |

to observe, but its presence is betrayed by a slight mottling of the etched surface, which cannot be found in a freshly quenched specimen, but is developed several months afterwards.[4a]

The damping peak does not appear in single crystals of iron if the alternating stresses are applied in ⟨111⟩ directions.[6] The reason for this is clear, since all the interstitial spaces in which the dissolved atoms lie are then distorted equally, and there is therefore no reason for the application of the stress to be accompanied by any change in the positions of the dissolved atoms.

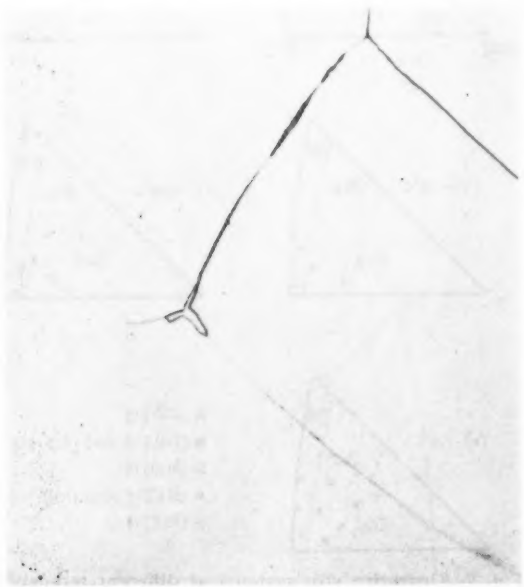


Fig. 4—Grain-boundary cementite in slowly cooled iron containing 0·0025%C × 1 500

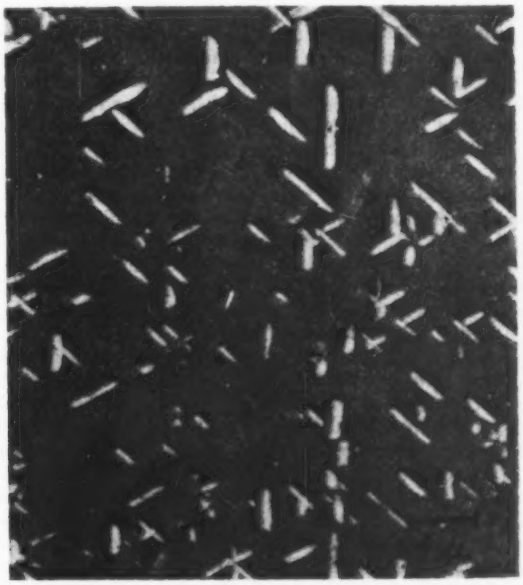


Fig. 5—Carbide precipitated at 200–250° C in the form of platelets in {100} planes

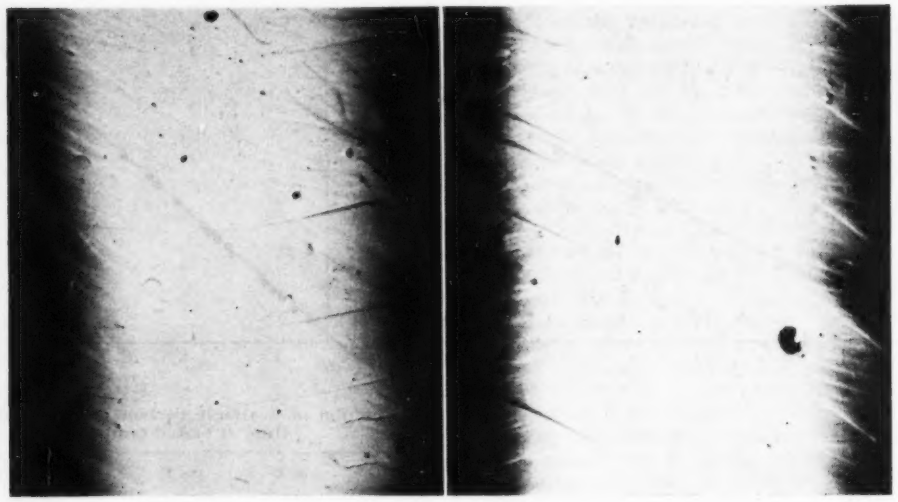



**Fig. 6—Slip lines in a pure iron single crystal: (*a*) viewed from the slip direction; (*b*) viewed at right-angles to the slip direction** × 80

**Plastic flow**

The most important early studies of the plastic flow of the iron crystal were those by Taylor and Elam and by Gough.[7] Taylor and Elam, using the large single crystals prepared by the method of Edwards and Pfeil,[8] demonstrated that the slip direction was always ⟨111⟩. They showed that the slip plane could vary widely, and considered that it was not always a simple crystallographic plane. The slip lines on the test-pieces were straight when viewed approximately at right-angles to the direction of slip, but wavy when viewed along the slip direction, and this they attributed to the ease with which the slip changed from one plane to another. They likened the movement to the glide in a bunch of hexagonal rods, in which the direction of shear is fixed, but the trace of the shear surface on a plane normal to this direction can take an irregular course.[7a] Gough investigated slip in alternating torsion, and although he agreed that the slip direction was always ⟨111⟩ he maintained that it occurred on the planes {110}, {112}, {123}. Whether or not slip in b.c.c. metals is crystallographic has been a source of mild controversy since then. Important contributions have been made by Barrett, Ansel and Mehl, Brick *et al.*, Chen and Maddin, and others.[9]

Some additional evidence on this has recently been obtained at the National Physical Laboratory.[10] The iron has been of somewhat higher purity than previous samples, and it has been found that provided the slip lines are examined at a very early stage of deformation, they appear smooth when viewed from any direction (Fig. 6) and correspond quite closely to traces of {110}, {112}, or {123} planes. At later stages of deformation the traces become wavy. A number of crystals of different orientations have been examined at the National Physical Laboratory and the observations concerning the slip plane are summarized in Fig. 7. When the direction of stress lies near the [$\bar{1}$11]–[011] edge of the unit triangle of the stereographic projection, there is a tendency for the slip plane to be [$\bar{2}$11], but as this direction moves

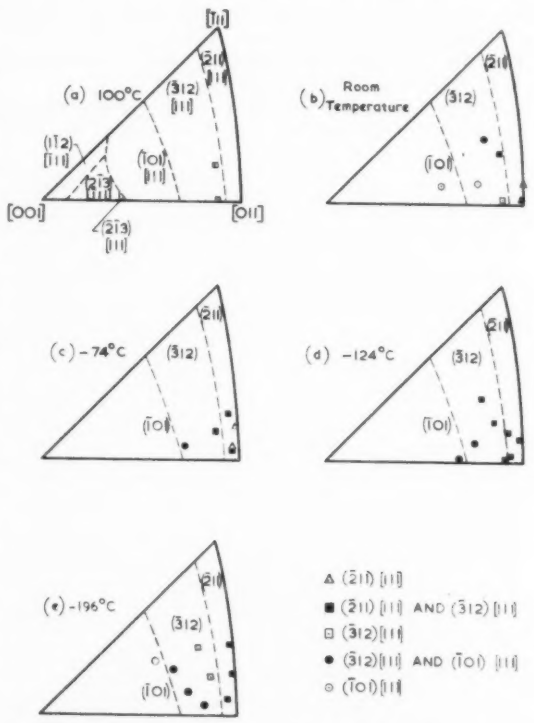



**Fig. 7—Operative slip systems at different temperatures. Dashed lines are the boundaries for equal critical shear stresses, the boundaries for the whole unit triangle being given at (*a*)**

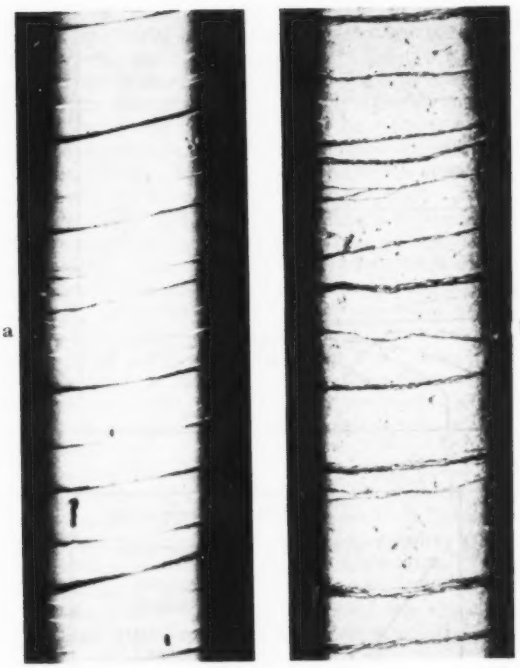



**Fig. 8—Appearance of slip bands in single crystals of iron containing (*a*) 0·05%N and (*b*) 0·016%C × 150**

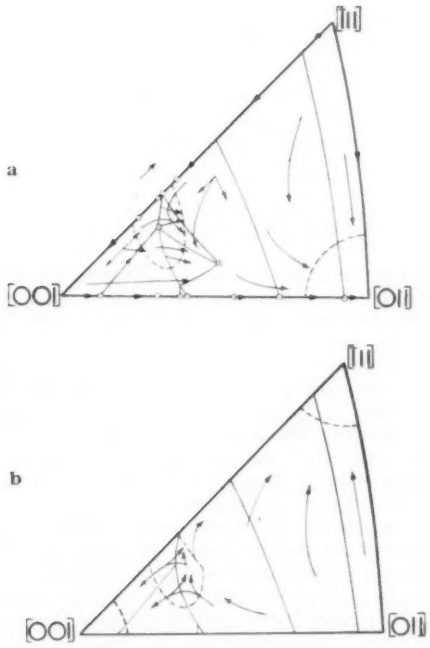



**Fig. 9—Rotations and final orientations in tension and compression: (*a*) tension; (*b*) compression**

**TABLE I**
**Comparison of yield stresses**

| Yield stress at room temperature, tons/in² | Authority |
|---|---|
| 2·4 | Paxton and Churchman[16a] |
| 3·6 | Steijn and Brick[9c] |
| 3·6 | Edmondson[12] |
| 3·8 | Cox, Horne, and Mehl[29] |
| 5·0 | Allen, Hopkins, and McLennan[10] |
| 5·6 | Vogel and Brick[9b] |

towards the [001] corner of the triangle one observes first a mixture of $(\bar{2}11)$ and $(\bar{3}12)$ slip, then $(\bar{3}12)$ slip alone, then a mixture of $(\bar{3}12)$ and $(\bar{1}01)$ slip, and finally $(\bar{1}01)$ slip alone. The procession of events is substantially the same at all temperatures. In Fig. 7*a* are shown the slip planes that would be found if slip occurred on either the $(\bar{1}01)$, the $(\bar{3}12)$, or the $(\bar{2}11)$ plane, and were determined by the magnitude of the shear stress on the plane resolved in the [111] direction, the critical shear stresses being the same for all types of plane. Broadly speaking, the observed slip follows the calculated pattern, but very often slip on two planes for which the resolved shear stresses are nearly equal is found where, on mathematical grounds alone, slip on one plane should be preferred. However, this is what really should be expected on dislocation theory, since slip is conceived as the movement of dislocation lines across the slip plane, and these lines are not all equally well placed for movement. It is very possible that the dislocation line most favourably placed for movement may occur on the plane having the lower resolved shear stress, and this would result in the kind of departure from ideal behaviour that is observed. On the other hand, Hempel, Kochendörfer, and Hillnhagen[11] have recently observed that under alternating stress, straight slip lines belonging to a single system occur when the stress direction lies in the middle of one of the single slip fields defined in Fig. 7*a* and slip lines belonging to two systems, or wavy lines, occur when it lies near a boundary.

This behaviour is not very different from that originally described by Taylor and Elam. Referring to Fig. 1*b*, it is seen that there are a number of {110}, {123}, and {112} planes containing any one slip direction and that slip on an arbitrary plane can be produced by a combination of short lengths of slip in neighbouring planes of these three types. Indeed, Fig. 1*f* shows that {123} slip can well be regarded as slip on two {011} planes, as has been suggested by Chen and Maddin,[9d] and this is very similar to the ' pencil glide ' proposed by Taylor and Elam.[7] It is not surprising that different investigators should have taken different views of this slip mechanism, for different samples of iron behave in characteristically different ways. For example, a single crystal of high-purity iron slips on numerous closely spaced planes, so that the mirror finish of the surface is maintained throughout the test (Fig. 6), but when nitrogen is added the slip planes are more widely spaced, and a greater amount of movement takes

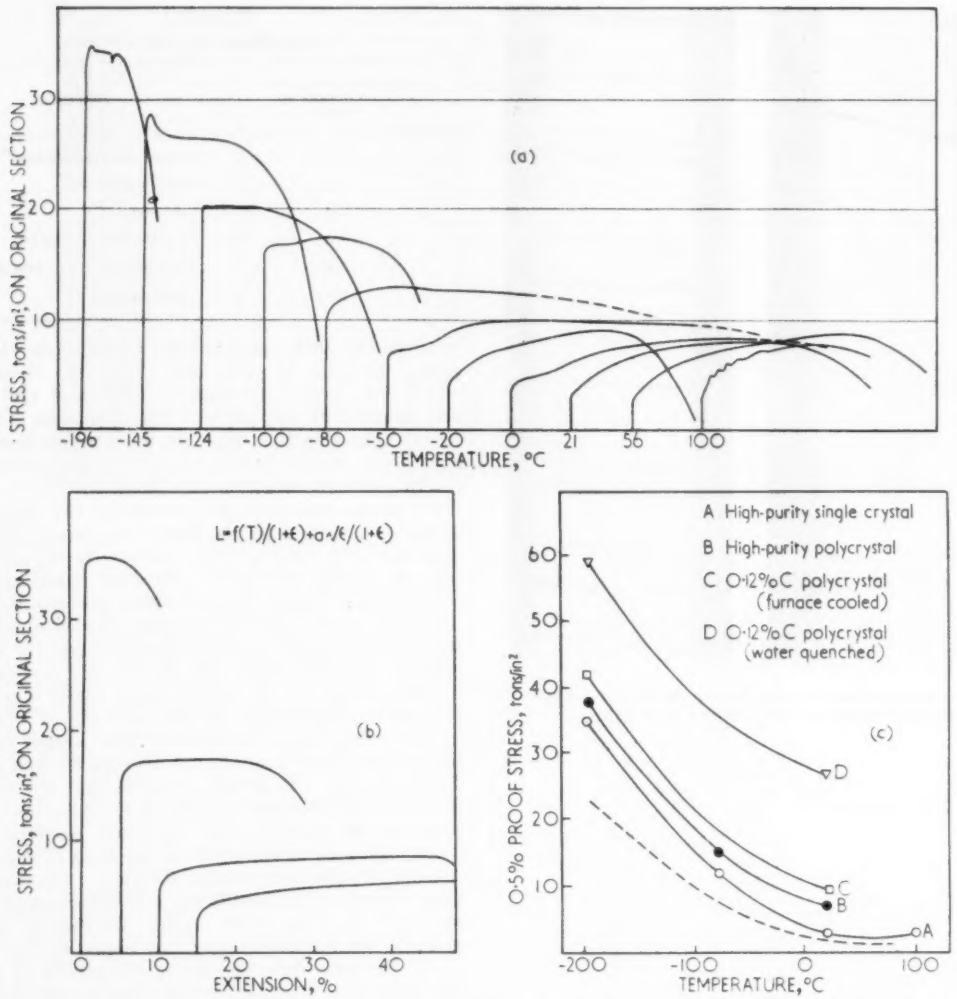



**Fig. 10—(*a*) Stress/extension curves of single crystals at various temperatures; (*b*) calculated stress/extension curves; (*c*) influence of purity and structure on the yield-stress temperature curve**

place on each of them, and when carbon is added, the traces become wavy (Fig. 8).[12] There is here a large and practically untouched field for investigation.

Plastic deformation in the manner described results in rotation of the crystal, and in polycrystalline masses continued deformation results in orientation textures. These textures have been studied by Calnan and Clews.[13] The rotations are relatively complicated (Fig. 9), but lead to the [011] direction coming into the direction of stress in the tensile case, and either the [001] direction or the [$\bar{1}$11] direction in the case of compression. In compression, crystals of rather similar orientation may rotate in different directions, and this, it might be imagined, would lead to very severe local stresses.

### The yield point

The stress at which plastic flow begins is highly dependent on the temperature and the purity of the iron and the method of making and testing the crystal. Table I contains some yield stresses at room temperature, derived from various authorities. The lowest figures are for carefully decarburized samples, and are about 2·4 tons/in², corresponding to a critical shear stress of about 1·2 tons/in². Less thoroughly decarburized samples are a little stronger. The two figures from the National Physical Laboratory are of interest, since they are averages of the results from a number of crystals made from the same iron, but grown by different methods.[10, 12]

The resistance to plastic deformation increases progressively as the temperature is lowered. Figure 10*a* shows a typical series of stress/elongation curves, relating to high-purity iron single crystals containing 0·0027%C and 0·0016%N,[12] which have been cooled slowly from the temperature at which they were grown and tested within a few days. The rise of load after the first yield is appreciable at all tempera-

tures above $-80°$ C, and between $+100°$ and $-80°$ C the total elongation at fracture tends to increase with falling temperature. Below $-80°$ C no appreciable increase of load occurs after the first yield. The elongation tends to be concentrated at one region in the specimen, and to decrease with falling temperature. At room temperature there is no sudden yield. At 0° to $-100°$ C there is generally, but not always, a suggestion of a yield point. Below $-124°$ C there is a definite maximum in the load/elongation curve, resulting in an upper and lower yield point. From 100° C the serrations in the load/elongation curve associated with strain ageing are observed. If several single crystals are tested at the same temperature, some will show a definite maximum at the yield point and others will not. The effect of temperature is therefore best expressed by saying that the probability of a definite maximum at the yield point increases progressively as the temperature is lowered.

Neglecting the behaviour at the yield point, the general pattern of the curves suggests that the iron crystal has an initial resistance to plastic deformation which rises rapidly with falling temperature, and a work-hardening capacity that is a function of the plastic strain. Assuming that the work hardening is proportional to the square root of the strain, in accordance with Taylor's original dislocation theory,[14] the load rises according to the equation

$$L = f(T)(1 + \epsilon) + a\sqrt{\epsilon}(1 + \epsilon)$$

(in which $T$ is the temperature and $\epsilon$ the strain) until the maximum load is reached, after which it falls away sharply as the specimen begins to constrict locally. Figure 10*b* shows the forms of load/elongation curve that would be expected on this basis, and reproduces the observed forms moderately well. The agreement, however, is only approximate, and the equation obviously needs to be refined in detail.

In Fig. 10*c*, curve *A*,[12] the proof stress of a typical single crystal is plotted against the temperature.*

* The proof stress of a single crystal of iron varies with orientation but is generally about twice the critical shear stress.

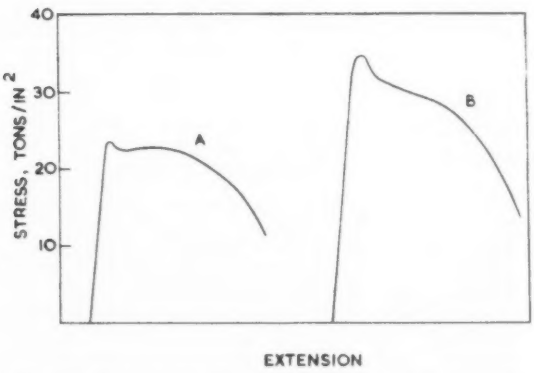



**Fig. 11—Stress extension curves at $-124°$ C of unaged and aged single crystals containing 0·0027%C: (A) immediately after cooling; (B) aged 2 years at room temperature**

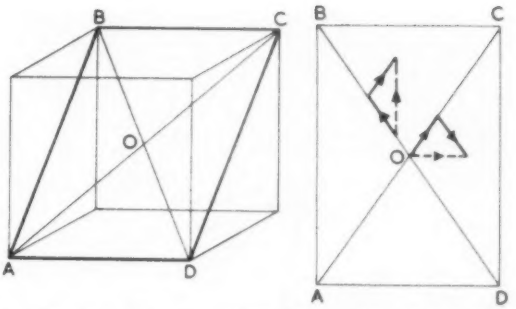


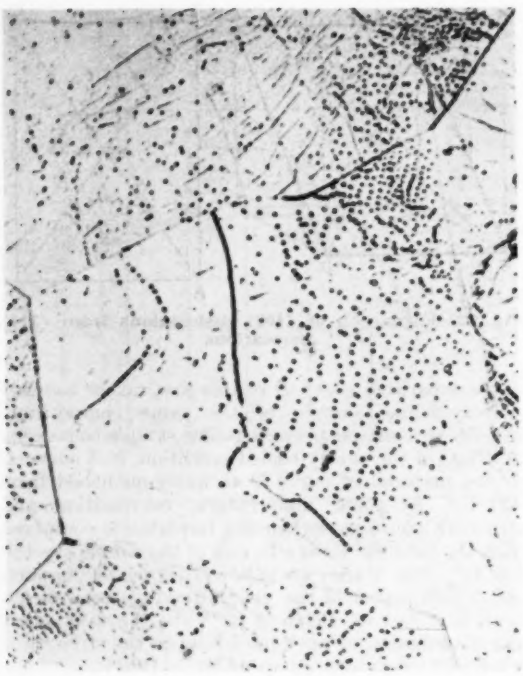

**Fig. 12—Formation of ⟨100⟩ dislocations from ⟨111⟩ dislocations**

For comparison, curve *B* relates to a rather coarsely polycrystalline sample of the same composition, curve *C* to a similar polycrystalline sample containing 0·12%C in the slowly cooled condition, and curve *D* to the material of curve *C* as water-quenched from 950° C.[4] At room temperature, composition and structure are of overwhelming importance compared with the intrinsic shear strength of the simple crystal, but at $-196°$ C they are relatively much less important. This aspect of the yield stress has been developed in detail by Petch *et al.*[15] In the same way, the contribution of work hardening to the strength is relatively much less important at $-196°$ C.

The material of curve *A* is not entirely free from carbon. For crystals containing less than 0·0005%C, the curve would run generally parallel to *A*, but appreciably lower, in the position of the dotted curve, which is derived from the work of Paxton and Churchman and the recent publication by Biggs and Pratt.[16] The effect of dissolved carbon on the proof stress of polycrystalline ferrite at 20° C has been determined by Wert.[3d] The last traces of carbon, of the order of two or three parts per million, have a considerable effect, but the proof stress continues to rise, as carbon is added, up to 0·005%C at least.

The yield-point phenomenon is influenced both by the way in which the crystal is grown and by the heat-treatment. In Fig. 10*a* no yield points are observed above 0° C, but in a recently reported series of experiments[10] crystals made from the same iron by another method showed clearly marked yield points at 100° C. Prolonged ageing at room temperature makes the yield point more prominent. This effect is well illustrated in tests made at $-124°$ C and reproduced in Fig. 11.[12] Curve *A* was obtained immediately after the crystal had been cooled rather slowly from 890° C, and curve *B* was obtained from a similar crystal tested after two years at room temperature. The time was sufficient to precipitate any carbon that remained in solution after the cooling from 890° C and the ageing resulted in a substantial increase of strength and an increase in the difference between the maximum of the curve and the stress at the lower yield point.

These observations have a bearing on the theory of the sharp yield point in iron. The phenomenon has been associated with the presence of carbon, and, ever since it was shown by Edwards and Pfeil that single crystals made from decarburized iron showed



**Figs. 13 and 14—Etch pit arrays in high-purity iron containing 0·086%P; normalized** × 750

no sharp yield at room temperature,[8] and later by several investigators that elimination of carbon and nitrogen, or additions to low-carbon steel of elements that would combine with carbon and nitrogen, would also eliminate the yield point.[17] The popular theory at that time was that carbides were precipitated in the grain boundary, and there exerted a resistance to slip which is suddenly overcome when the sharp yield occurs,[18] but Edwards, Phillips, and Liu[17c] concluded that the resistance was due to the precipitation of carbides upon certain planes of slip within the crystal. These views lost ground when the smallness of the amount of carbon or nitrogen necessary to produce the yield was realized, for the necessary quantity seemed far lower than the solid solubility of these elements in ferrite, as it was then thought to be, and lower than the the amount necessary to cause visible separation of cementite in the grain boundaries. The alternative theory, that dissolved carbon or nitrogen segregated to the dislocations in the crystal and there ' anchored ' them, was put forward and forcefully argued by Cottrell[19] and has been widely accepted. But this theory involves the conclusion that the increase of yield stress arising from this cause reaches a maximum when the quantity of dissolved carbon or nitrogen is sufficient to anchor all the dislocations, and does not of itself readily explain the rise of yield stress that occurs when the quantity of interstitial solute is greater than this very low value, nor the further rise that occurs when the conditions are such as to favour precipitation of the solute. It seems likely, therefore, that any final theory of the yield point will have to take into account

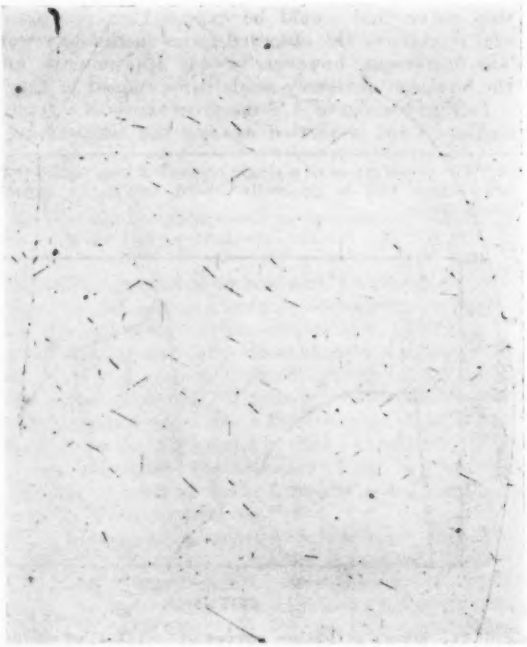


**Fig. 15—Etch pit arrays in high-purity iron containing 0·05%P; furnace-cooled from 950° C** × 150

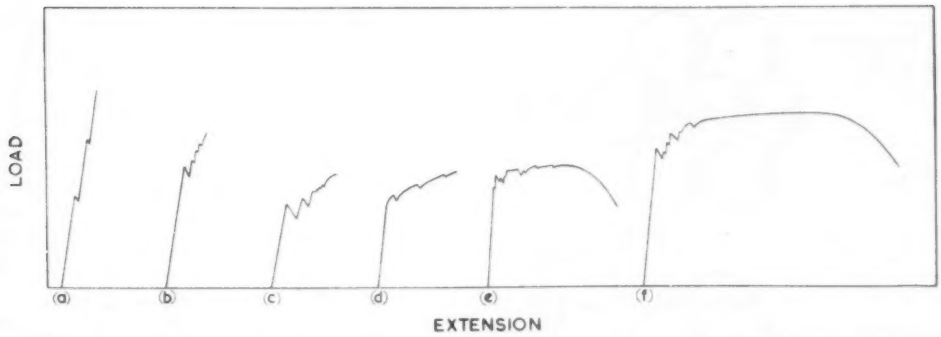



**Fig. 16—Load/extension curves of iron crystals showing twinning with or without plastic deformation**

additionally the well known anchoring of dislocations by finely precipitated particles, and that it will be necessary to stipulate that the size and number of these particles influence both the stress at which the dislocations break away from their initial positions and the stress at which they subsequently move through the crystal. Since there is evidence that precipitated particles grow at dislocations,[20] there may be a continuous series of stages between a vaguely defined ' atmosphere ' of dissolved carbon or nitrogen atoms in the neighbourhood of a dislocation and the appearance of a perfectly formed crystal of cementite or iron nitride at the site of a pre-existing dislocation.

**Dislocations in iron crystals**

The considerations in the preceding paragraph lead to an interest in the nature and distribution of the possible dislocations of an iron crystal. The theory of the dislocations of a b.c.c. lattice is less advanced than that of the dislocations of a f.c.c. lattice, but at least two types of dislocation are recognized. The first is a dislocation having a Burgers displacement vector of $a/2$ in a $\langle 111 \rangle$ direction. This must be a fairly mobile dislocation, since slip in iron is in $\langle 111 \rangle$ directions. The second[21] is made by combining two such dislocations having Burgers vectors in different directions, which can be done with a reduction of strain energy, provided that the signs of the two displacements are appropriate. In Fig. 12, $ABCD$ is the (110) plane containing two $\langle 111 \rangle$ directions $AC$ and $BD$. Dislocations with vectors in the directions $OC$ and $OD$ will combine according to the reaction

$$\frac{a}{2}[\bar{1}11] + \frac{a}{2}[1\bar{1}1] = a[001]$$

which results in a decrease of strain energy, whilst those with vectors in the directions $OC$ and $OB$ do not combine because the reaction

$$\frac{a}{2}[\bar{1}11] + \frac{a}{2}[\bar{1}1\bar{1}] = a[\bar{1}10]$$

is attended by an increase of strain energy. The new dislocation formed according to the first equation has a Burgers vector in the direction $AD$, parallel to the edge of the unit cube and, since this is not a favoured direction for slip in the iron crystal, this type of dislocation is probably immobile. Because of its large Burgers vector, the strain energy associated with this dislocation is relatively large, and the cavity on the tension side of the dislocation line is also rather large.

Only the first of these dislocations has hitherto been definitely observed. This was achieved by study of the etch pits produced in a single crystal of iron containing 3·5%Si using a special electropolishing and etching procedure and described in an elegant paper by Hibbard and Dunn.[22] The crystal was prepared in such a way that one side of the rectangular bar was the $(\bar{2}11)$ plane, and the $(01\bar{1})$ plane, which is a preferred slip plane in silicon iron, lay at 45° to the axis of the bar. The crystal was bent around the $[\bar{2}11]$ axis, and the etch pits appeared on the $(\bar{2}11)$ surface in lines running at 45° to the axis of the bar in exactly the way that would be expected if they represented the ends of dislocation lines which had travelled across the specimen in the $(01\bar{1})$ plane. The number of etch pits agreed with that expected from dislocation theory, and when the crystal was annealed the etch pits rearranged themselves into lines perpendicular to the original slip lines in precisely the way demanded by the theory of polygonization. Having established that these pits did indeed represent the positions of dislocation lines, it was possible by counting them to show that the original undeformed crystals had contained about $2 \times 10^6$ dislocations per cm$^2$, and that after deformation $10^7$–$10^8$ dislocations per cm$^2$ were present.

Dislocation etch pits of this kind are not so easy to develop in unalloyed iron as in the 3·5%Si alloy. The first success appears to have been reported by Boswell,[23] and it may be of interest to describe some of the experiences at the National Physical Laboratory.[24] They are not found in the purest iron we have prepared, but they are readily produced if about 0·006%C is present, and best if the material is cooled from a high temperature at an appropriate rate. They are very nicely developed if nitrogen, oxygen, phosphorus, silicon, or molybdenum is present, but curiously enough, not when chromium is added, and our impressions confirm the opinion that they are due to the concentration of impurities at regions of local strain and that the conditions for their best development are rather critical. Consequently when, for example, in a very pure iron the pits fail to develop, it is not concluded that dislocation arrays are not present, but that the conditions for displaying them have not been achieved.

The distribution of the etch pits, even in normalized iron, is extraordinarily irregular. Within one crystal there may be comparatively clear patches, irregular

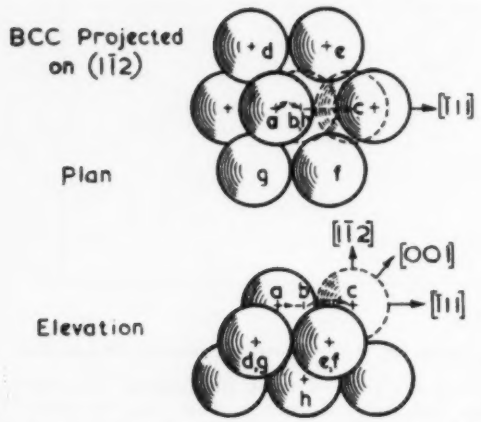


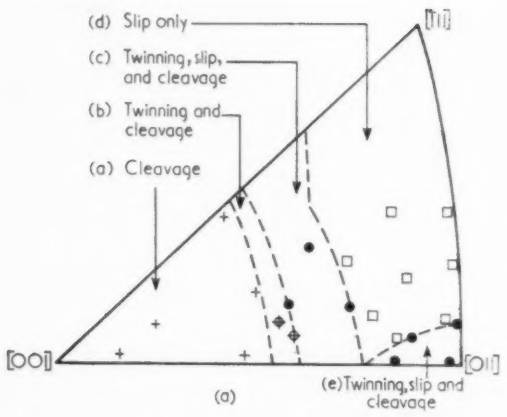

**Fig. 17—Movements of atoms on twinning on {112} planes in b.c.c. lattices**

lines which on high magnification often prove to be lines of closely spaced pits, and closely packed arrays in which the density of pits may rise to $10^8$ per $cm^2$ or more (Fig. 13). The array in Fig. 14 closely resembles the arrays in Hibbard and Dunn's deformed and annealed specimens, and probably has the same origin, i.e. it probably represents a region that has at some time been locally deformed and where the resultant dislocations have persisted in spite of the heat treatment. In slowly cooled specimens arrays of the kind illustrated in Fig. 15 can often be found and are frequently resolvable into rows of pits. They probably represent regions where precipitation is specially favoured. In the crystal shown in Fig. 15, six directions of these arrays are found, so that they could conceivably represent arrays of dislocations lying in {110} planes. One is led by these observations to think that even annealed crystals are not strain-free, but that local concentrations of strain energy, accompanying local concentrations of dislocation lines, are present.

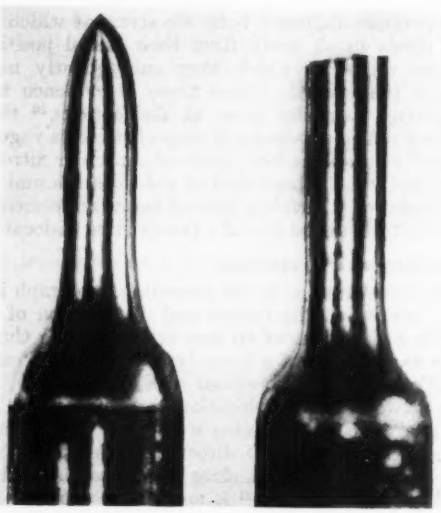
### Twinning

Before proceeding to the phenomena of fracture in single crystals, it is necessary to mention the formation of mechanical twins. Mechanical twins are produced during the deformation of iron, particularly when the temperature of deformation is low or the speed of deformation is high, as in the attack of mild steels by explosives. Their occurrence in single crystals was extensively studied by Pfeil,[25] who observed, among other things, that mechanical twins were more difficult to produce if the crystal had been previously deformed. In the testing of single crystals, their formation is accompanied by a quite audible 'cry' and by the appearance of a sudden step on the load/elongation curve, showing that the relief of stress is much faster than the application of load and, indeed, is almost instantaneous.

Although associated with the conditions leading to brittle fracture in iron, there is no evidence that the occurrence of twinning is necessarily a part of the fracture process. In Fig. 16 a variety of load/extension curves of single crystals are shown.[12] In




b



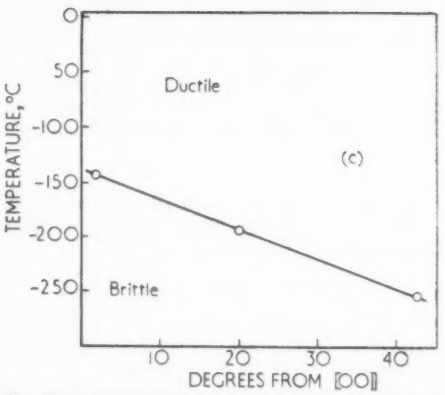



**Fig. 18—Fracture of iron crystals: (*a*) influence of orientation at − 196° C; (*b*) typical broken test-pieces (× 15); (*c*) influence of temperature and orientation**

curve *a* repeated twinning occurs well before brittle fracture with no trace of other plastic deformation. In curves *b–d* twinning is accompanied by greater or

lesser amounts of normal plastic deformation and the ultimate fracture is sudden and accompanied by no local necking. In curves *e* and *f* repeated twinning is followed by a long period of normal plastic deformation leading to a ductile fracture with extensive local constriction. The twinning process thus appears to be imposed upon the normal processes of plastic deformation and fracture, without affecting them materially.

Twinning occurs on {112} planes in ⟨111⟩ directions and involves a shear movement which differs from {112} slip in that the relative movements of atoms on adjacent planes are less. Whereas in slip an atom in the position *a* (Fig. 17)[26] moves to the position *c*, and may make a similar movement repeatedly, in twinning only the single movement from position *a* to position *b* occurs. Consideration of the geometry of the lattice suggests that the twinning movement should occur in one direction only, so that the twinning systems that operate in tension should not operate when the stress is reversed to become compressive. If it is assumed that a critical shear-stress criterion can be applied to twinning, it appears that whilst the critical stresses for twinning are for normal rates of strain above those for slip at high temperature, they are rather lower at − 196° C, so that at this temperature there are some orientations for which slip occurs without twinning and others for which twinning precedes slip. There is little doubt that the production of twins at high rates of strain is connected with the great speed at which twinning produces large strains. Strain due to slip commonly develops more slowly, and the stress must be raised considerably to produce rapid strain. Consequently it is not surprising that the quick deformations that accompany fracture often result in the formation of twins near the surface of the fracture. These twins have recently been studied in detail by Berry.[27]

### Fracture

At high temperatures single crystals of iron fail by the familiar process of slip on two planes, leading to the formation of a wedge-shaped reduced section which, if the metal is pure, becomes so thin that the fracture has a chisel edge. The breaking stress can be roughly estimated by measuring the load and the reduced area periodically during the test and extrapolating the true-stress/true-strain curve to the point corresponding to the final fracture strain. In an experiment of this kind conducted on a single crystal at room temperature, the highest stress observed before fracture was 15 tons/in², and the true stress at fracture was probably two or three times higher, whilst in a similar experiment[28] made at − 196° C the highest observed stress was 84 tons/in² and the probable fracture stress not less than 130 tons/in²,[28].

When the temperature is lowered, the behaviour changes. At a rather vaguely defined temperature, which depends on the sample, some of the crystals are found to break by a sudden cleavage across a {100} plane without any apparent prior deformation, whilst others extend and draw down to a chisel edge exactly as at high temperature. The critical temperature was near − 196° C in experiments reported by the National Physical Laboratory[10] and by Cox, Horne, and Mehl[29] and around − 110° C in those reported by Steijn and Brick.[9c] The behaviour depends on the orientation of the crystal in relation to the direction of stress and is relatively complicated. Figure 18*a* summarizes the behaviour of the samples examined at the NPL. In the region *d* the crystals are entirely ductile, undergo slip processes without any trace of cleavage, and draw down to a sharp wedge. In regions *c* and *e* twinning is followed by slip, and then by a cleavage fracture. In region *b* twinning is followed by cleavage, without any intervening slip, and in region *a* there is only cleavage. The orientations for which only cleavage occurs are those for which a {100} plane is almost normal to the direction of stress, and therefore those for which the resolved stress across the cleavage plane is greatest. The resolved stress at which fracture occurs is rather irregular, but averages about 30 tons/in². Figure 18*b* illustrates typical broken test-pieces. At − 196° C the boundary between essentially ductile and essentially brittle behaviour lies about 21° away from the [001] direction (Fig. 18*a*). This angle apparently varies very sharply with temperature, for at − 145° C only specimens very close to the [001] orientation have shown any trace of brittleness, whilst at − 253° C all specimens are brittle except some with orientations very close to the [011] direction, which twin a little before breaking. The effect of temperature on the behaviour must therefore be approximately represented by Fig. 18*c*. The average fracture stress is about 40 tons/in² at − 253° C, which is appreciably higher than that observed at − 196° C.

The influence of orientation on cleavage can also be observed at room temperature, if the crystal is subjected to shock. In 1905 Osmond and Frémont[30] examined some large crystals found in a piece of oxidized and decarburized steel rail that had been used for 15 years in the bracing of a furnace. An unnotched test-piece cut with its axis normal to a {100} plane broke in an impact test without recorded deformation, by cleavage, without appreciable absorption of energy. A similar piece with a {100} plane inclined at 30° to the axis bent with considerable absorption of energy and did not break. A specimen with a saw cut, and a {100} plane normal to the axis, bent without breaking and with much deformation at the base of the saw cut, when the load was applied slowly.

Stead had observed a similar behaviour seven years before[31] in coarse-grained decarburized mild steel sheet, and Pfeil's single crystals could be broken by a suitably directed sharp blow at room temperature; however, it has not yet proved possible to cleave at room temperature the high-purity iron single crystals produced at the NPL.

### The theory of fracture

The fracture behaviour of single crystals is somewhat disconcerting, for recent theories of fracture[21, 32] have proceeded from the assumption that a certain amount of plastic deformation is necessary before fracture, and generally that fracture is initiated at the highly stressed areas produced at the end of a

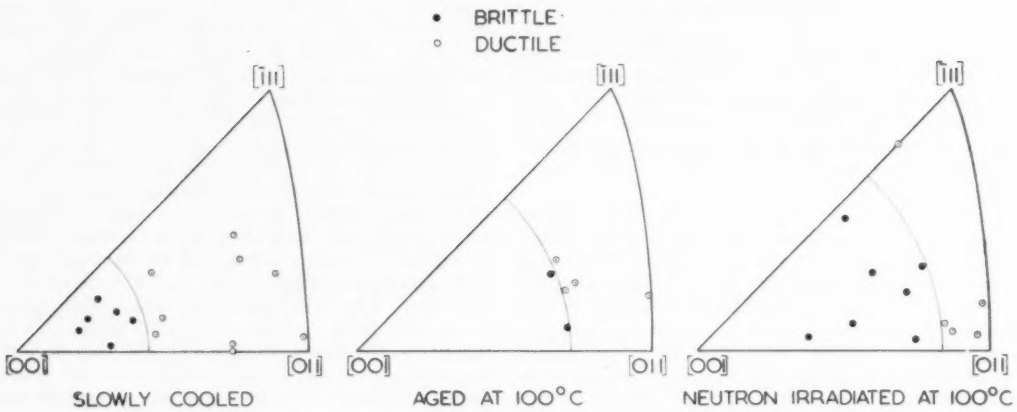



**Fig. 19—Fracture of single crystals of iron at — 196° C: (*a*) slowly cooled from 890° C; (*b*) slowly cooled and aged 5 days at 100° C; (*c*) slowly cooled and irradiated with neutrons ($8 \times 10^{17}$ n/cm²) for 5 days at 100° C**

slip line when the slip dislocations are arrested at a boundary. But when the suitably orientated crystals cleave there is no sign of prior deformation, and no boundary to arrest the dislocations. In order to maintain the theory one might assume that only a microscopic amount of slip is necessary, and that this could have occurred at some point of stress concentration due to non-axial loading, or to some surface imperfection, or to some internal flaw. It has, of course, been known for a very long time that by sufficiently sensitive methods minute plastic deformations can often be detected at stresses well below the elastic limit as conventionally observed.[33] But this explanation is unsatisfactory, for it is necessary to explain why a crystal tilted in one direction is perfectly ductile, whilst one tilted in another direction is quite brittle, and the assumption of non-axial loading or local flaws will not do this. Indeed, it will be found that no explanation is possible that does not assume at some stage of the argument that the stress at which cleavage occurs, and the stresses at which twinning or slip occurs are under these particular conditions of test quite close to one another and are so affected by the orientation of the crystal that in one orientation the condition for cleavage is reached before the condition for general yield, and in another orientation the reverse is the case. The evidence is that the occurrence of slip obstructs fracture rather than promotes it, for when the stress in the reduced section of a ductile specimen has reached 84 tons/in², the crystal has re-orientated itself until a {100} plane is at 45° to the direction of stress, and the resolved tensile stress across the cleavage plane is 42 tons/in² compared with a cleavage stress of about 30 tons/in² for the undistorted crystal.

The attempt has been made to study the relation between the yield stress and fracture stress of the crystal at — 196° C by hardening the crystal in various ways and observing the effect on both properties. The work is very tedious because the behaviour of the crystals is not very reproducible, and it is necessary to obtain averages from a large number of samples. The occurrence of twinning in many specimens also complicates the interpretation of the data. Nevertheless, data have been obtained for crystals hardened by dissolved interstitial atoms, by the presence of small precipitated particles, and by ageing at 100° C with or without exposure to neutron irradiation. The results, which are due to Edmondson,[12] are given in Fig. 19 and Table II in terms of the average resolved tensile stress across the (001) plane at fracture, and the orientation at which the crystals change from an essentially ductile to an essentially brittle behaviour. It is seen that a change of the critical shear stress is not in general accompanied by an equal change of resolved fracture stress, and as a result the critical angle changes. Apparently ageing and radiation damage both raise the critical shear stress without significantly affecting the fracture stress, whilst nitrogen increases both the critical shear stress and the fracture stress, but not in the same degree. In Table II, except for the case in which there are only five observations in all, the changes of fracture stress and shear stress are statistically significant. Not enough experiments have yet been made to ensure that the changes in the ratio of fracture stress to shear stress, taken by themselves, are significant, but taken in conjunction with the changes of critical angle illustrated in Fig. 19, which would not occur if the changes of ratio were due to chance, it seems fairly certain that the shear stress and the resolved

**TABLE II**

**Stresses in various crystals (after Edmondson[12])**

| Description | Resolved shear stress (av.), tons/in² | Resolved cleavage stress (av.), tons/in² | Ratio | Transition angle |
|---|---|---|---|---|
| Iron: slowly cooled | 18·8 | 26·6 | 1·41 | 21° |
| Iron: aged at 100° C | 19·8 | 25·7 | 1·30 | 34° |
| Iron: neutron-irradiated at 100° C | 23·1 | 28·5 | 1·23 | 38° |
| 0·05% nitrogen: quenched | 24·9 | 31·5 | 1·26 | 39° |
| 0·05% nitrogen: quenched and aged at 100° C | 35·0 | 40·4 | 1·15 | 41° |

fracture stress are controlled by independent factors. A recent paper by Biggs and Pratt[16b] describes a similar change in the critical angle produced by the addition of carbon.

Neither the critical shear stress nor the resolved fracture stress measured in these experiments represents the best possible properties of an iron crystal, for it is well known that very tiny crystals of pure iron made in the form of ' whiskers ' of a few microns thickness by reducing iron bromide with hydrogen are very strong even at room temperature and, if sufficiently thin, neither yield nor break below a stress of some hundreds of tons per square inch.[34] It is therefore necessary to conclude that the more massive crystals are imperfect, but that since there is no direct relation between the yield stress and the cleavage stress, the imperfections that lead to fracture are not produced by yielding, but are already in the crystal.

In considering the possible nature of these imperfections, a very simple experiment with single-crystal whiskers of iron is instructive. Such a crystal may be bent elastically to a very high curvature, such that the calculated stress on the outer fibres is of the order of 50–100 tons/in². But when so bent it is not in its condition of least energy: after a while it can occasionally happen that it will kink at one point, clearly because the energy consumed in the production and movement of dislocations at the kink is less than the elastic energy released by the straightening of the remainder of the whisker. The process is a means of dissipation of elastic strain energy. The production and spread of cracks is another process whereby elastic strain energy is dissipated. In a long chain of atoms under tensile strain, the elastic condition whereby each interatomic distance is increased by the same amount is not the condition of least energy, since, for every length of the chain there is a stress at which the energy is less if by any chance one link is broken, and the remainder are allowed to relax.

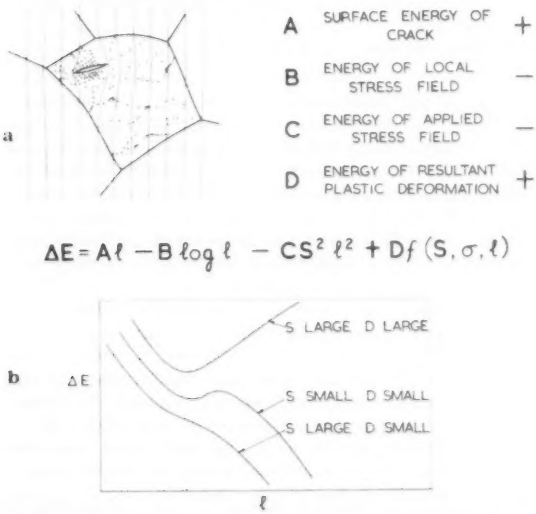

a
A SURFACE ENERGY OF CRACK +
B ENERGY OF LOCAL STRESS FIELD −
C ENERGY OF APPLIED STRESS FIELD −
D ENERGY OF RESULTANT PLASTIC DEFORMATION +
$\Delta E = A\ell - B\ell\log\ell - CS^2\ell^2 + Df(S, \sigma, \ell)$
b
ΔE
S LARGE D LARGE
S SMALL D SMALL
S LARGE D SMALL
ℓ

**Fig. 20—(*a*) Diagram illustrating formation of a crack in a region of stress concentration; (*b*) energy as a function of crack length**

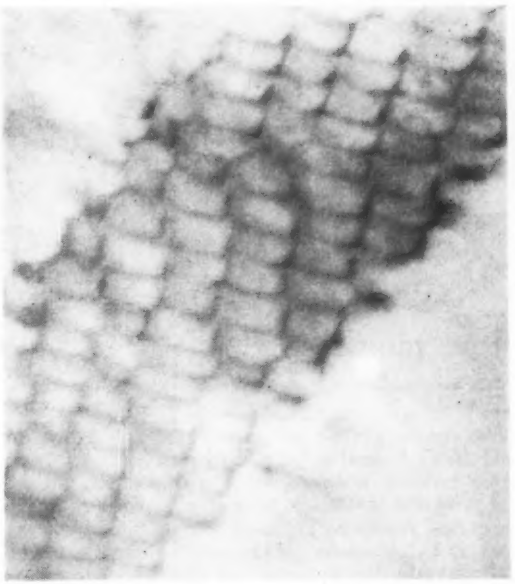


**Fig. 21—Dislocation network in annealed ferrite × 100 000**

For an infinitely long chain this stress approaches zero, so that for an infinitely long crystal there is no stress below which the danger of fracture does not exist. Griffith,[35] indeed, began his well known treatment of the rupture of solids with the statement that "the equilibrium position (of an elastically deformed solid body) . . . must be one in which the rupture of the solid has occurred, if the system can pass from the unbroken to the broken condition by a process involving a continuous decrease in potential energy."

The reason why a large crystal displays tensile strength is that in a large cross-linked array of chains, such as a crystal may be supposed to be, there is no way in which parting can begin that does not involve an initial increase in energy. If a small crack appears in a stressed body, the surface energy of the two new surfaces must be supplied at the expense of the elastic energy released in the surrounding material. But the release of elastic energy is not as great as might be expected on account of the high concentrations of stress produced at the edges of the crack, and the process as a whole does not result in a reduction of energy unless

$$S > K\sqrt{\frac{E\gamma}{\ell}}$$

where $S$ = stress
$E$ = Young's modulus
$\gamma$ = surface tension
$\ell$ = crack length
$K$ = constant of order unity.

Plastic deformation and fracture are thus two competitive processes for dissipating the elastic energy of the body. Both are subject to the requirement that (apart from the supply of activation energy from the thermal energy of the crystal) they cannot take place unless all stages of the process result in a reduction of energy. But if the circumstances are such that

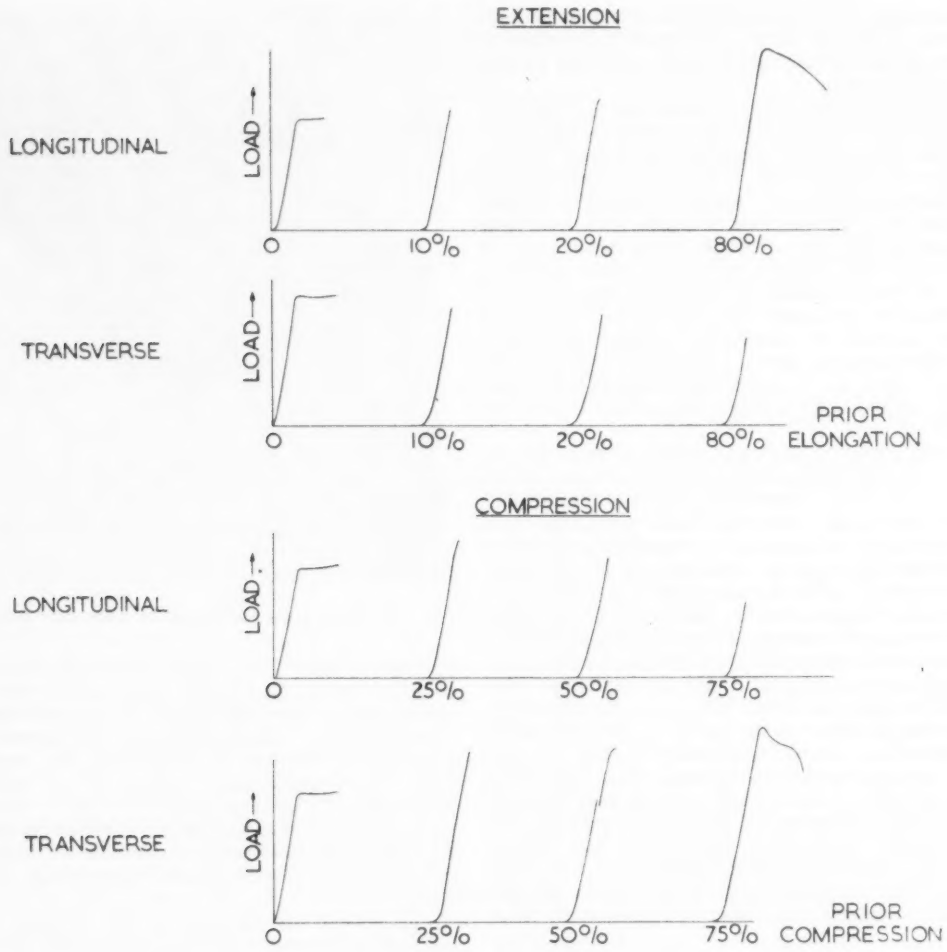



**Fig. 22—Load extension curves at −196° C of cold-worked iron**

energy can be reduced, both can and will take place.

Now consider Fig. 20*a*, which illustrates a material such as we know the iron crystal to be, with complex groups of dislocations arranged in a variety of loosely and closely packed patterns. The dislocations are all surrounded by stress fields, and where the dislocations are close together, the associated strain energy is high. When a stress is applied, the level of strain energy will be increased throughout. If the dislocations are loosely bound, some may move and release strain energy in that way, but if they are tightly bound, the possibility that the strains will be relieved by the formation of a crack exists. If a crack forms, it will probably do so in one of the highly stressed regions, and only if its formation can relieve the strain energy of the associated dislocation group. Its formation will result in a local redistribution of stress, and this may result in the movement of neighbouring dislocations that had previously been unmoved. There are thus four elements in the energy balance: (1) the surface energy of the crack, (2) the release of the strain energy of the local internal-stress field, (3) the release of the strain energy of the applied-stress field, (4) the energy required by the resultant plastic deformations. It is clear that when the crack is small, and of the same size as the average distance between dislocations, the surface energy of the crack and the strain energy of the local field exert an important influence, but when the crack is large, these factors are overwhelmed by the third and fourth factors.

It is impossible within the available space to develop these ideas at all fully, but two points may usefully be made. The first term is proportional to the length of the crack and independent of the applied stress or the condition of the metal, the second is proportional to the logarithm of the length of the crack and independent of the applied stress, the third is proportional to the applied stress and to the square of the length of the crack, and the fourth is proportional to the length of the crack and is strongly influenced by both the yield stress of the material and the applied stress. The energy change $\Delta E$ is

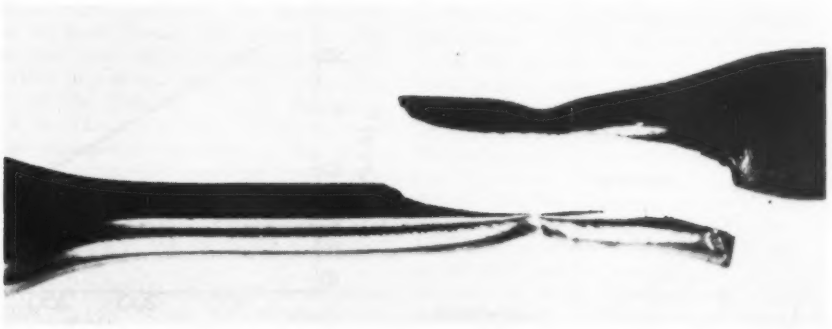


**Fig. 23—Longitudinal cleavage fracture in crystal of iron containing 0·05%N × 10**

therefore given by the following relationship

$$\Delta E = Al - B \log l - CSl^2 + Df(S,\sigma,l)$$

where $l$ = length of crack
$S$ = applied stress
$\sigma$ = yield point of material
$A$, $B$, $C$, $D$ are constants.

In this expression the first three terms are similar to those proposed by Stroh.[32f] If the term in $D$ is small, such a function when plotted against the crack length passes through a minimum if $S$ is below a certain value, which implies that any incipient cracks formed at the strain centres do not grow beyond a certain length, but as $S$ is raised a point is reached at which the minimum vanishes, and at this stress the crack grows indefinitely. On the other hand, if the $D$ term is large, as when large numbers of mobile dislocations are present, and the applied stress approaches the yield point, the minimum of the function does not vanish, and unless an initial crack of some length is present the material will not break but will deform plastically until it has work-hardened sufficiently to render the $D$ term negligible (Fig. 20*b*). The critical importance of the presence of a minimum in the energy/length curve is believed to be the cause of the sharpness of the transition from ductile to brittle behaviour, and the influence of orientation in a single crystal may be attributable to its influence on the third and fourth terms of the energy expression.

Secondly, in any system in which the elastic strain energy contained in a limited volume ($V$) of highly stressed metal supplies the surface energy necessary to produce a crack across that volume, the strain energy is proportional to $\frac{1}{2}ESV^{\frac{2}{3}} \times SV^{\frac{2}{3}}$ whereas the surface energy is proportional to $\gamma V^{\frac{2}{3}}$, where $V$ = volume of stressed element, $S$ = stress, $E$ = Young's modulus, and $\gamma$ = surface energy. Fracture occurs when $\frac{1}{2}ES^2V = \gamma V^{\frac{2}{3}}$ and $S = \sqrt{\left(\frac{2\gamma}{E} \cdot \frac{1}{V^{\frac{1}{3}}}\right)}$. That is $S \propto (1/\sqrt{l})$, where $l$ is some linear dimension characteristic of the microstructure. This is probably the origin of the commonly observed linear relation between the breaking stress and the square root of the grain diameter.[32d]

In view of these considerations, it has become desirable to know whether highly strained regions such as have been assumed are truly present in annealed iron, and whether they can be influenced in such a way as to alter the cleavage stress of the crystal. I am indebted to my colleague Dr McLean for Fig. 21, which is a transmission electron micrograph at a magnification of × 100 000 of a sample of annealed high-purity iron. This clearly shows a complex array of dislocations, consisting of two sets of parallel dislocations which have formed short lengths of dislocations in a third direction at their points of intersection. In other nets such as this it has been possible to prove by determining the orientation that dislocations with Burgers vectors parallel to ⟨100⟩ directions are present. These will have larger vectors than the ⟨111⟩ slip dislocations and will probably be immobile, giving stability to the whole system. They will also involve the presence of tensile stresses, mainly in ⟨100⟩ directions, which will be considerable in a region of the same order of magnitude as the size of the array.

If such an array were submitted to a shear stress encouraging one of the sets of slip dislocations to move, it would exhibit considerable resistance, but if the stress were high enough one of the dislocations might be expected to escape from time to time, and in due course the whole array would decompose. On the other hand, such a stress would probably cause considerable slip in other parts of the crystal, and in any place where heavy slip was occurring in two different directions there would be the possibility of forming new arrays of a similar kind. The new arrays would, however, be differently arranged from those that had been decomposed, for their Burgers vectors would be related to the operative slip directions, and consequently to the nature of the strain.

It is not easy to forecast from first principles the nature of the arrays that would be formed, and therefore the course has been followed[28] of taking a fairly pure sample of polycrystalline iron, subjecting it to heavy deformation, both by axially compressing a cylinder and by cold-rolling a rod, and studying the influence of the deformation on the fracture stress at − 196° C of test-pieces taken in different directions. The results are shown in Fig. 22. In both cases the fracture stress is progressively lowered in the direction in which the dimensions have been reduced. In the elongated direction, a material that was originally rather brittle has become ductile, and its resistance to fracture, far from being reduced by the plastic

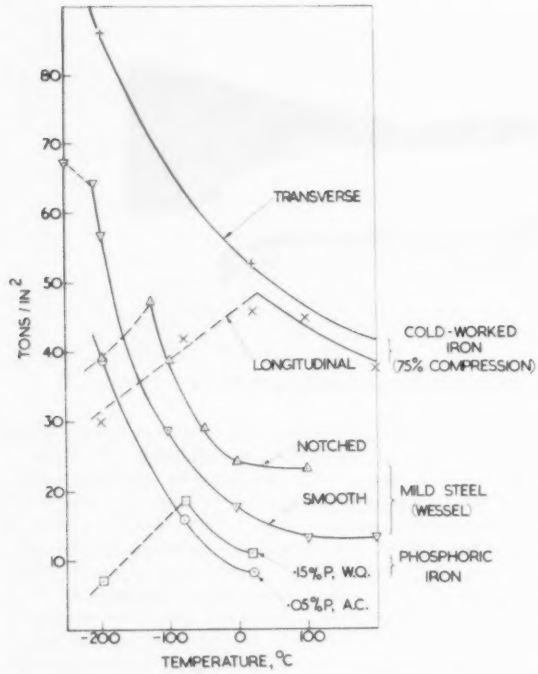



**Fig. 24—Effect of temperature on proof and fracture stresses of various polycrystalline samples as indicated**

deformation, is very much increased. In the reduced direction the fracture stress has become so low in relation to the yield stress of the material that there is no question of fracture being initiated by general yield. Careful microscopic examination shows these specimens to be free from any trace of flaws, and after being annealed their properties are quite normal.

It may be concluded from these observations that the dislocation arrays produced by plastic deformation are either such as to produce tensile stresses normal to the direction of elongation, or that the sizes and shapes of the arrays are such as to favour fracture in this direction. In the case of the rolled bar it is easy to imagine that a high transverse tension might have been produced, for the orientation assumed by the crystals is such that slip takes place in directions which would produce [100] dislocations having Burgers vectors in the direction at right-angles to the rolling direction. Moreover, ductile single crystals tested in tension sometimes split longitudinally in a way that suggests that weakness has been produced at right-angles to the line of intersection of the slip planes (Fig. 23).[12] But the compression case is not so easy, for the final orientations give no reason for supposing that the [100] dislocations would be specially well placed to give tensions in the required direction. It is probable, therefore, that the shapes and dimensions of the arrays are principally concerned, and that no explanation will arise until these have been examined in detail.

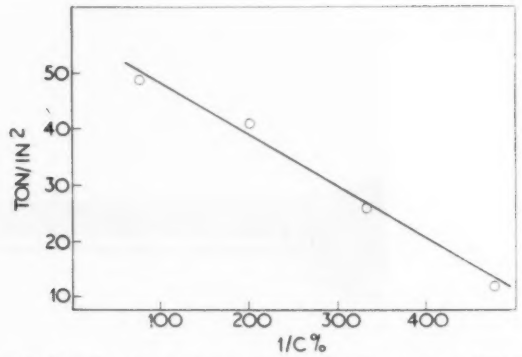



**Fig. 25—Fracture stress of 1%Ni alloys at — 196° C plotted against the reciprocal of the carbon content**

**Effect of temperature on fracture stress**

The effect of temperature on the behaviour of the cold-worked specimens will now be considered. In Fig. 24 the full curves show the proof stresses of samples that underwent a general yield before fracture, and the broken curves the fracture stresses of those that broke without appreciable deformation. The two curves at the top of the diagram relate to the iron that was cold-worked by compression. The proof stresses of the transverse test-pieces, which were ductile at all temperatures, increased as the testing temperature fell. The longitudinal specimens, which were weak and brittle at — 196° C, were ductile at 100° C and above, and their proof stress rose with falling temperature. They became brittle at about room temperature, and in the brittle condition their fracture stress decreased sharply as the temperature decreased. It is clear that at — 196°C the breaking stress was well below the stress at which general yield could have been expected to take place.

The other curves in this diagram contrast the behaviour of specimens made tough or brittle in different ways. An iron containing 0·05%P when air-cooled from 950° C is ductile at all temperatures above — 196° C, though at — 196° C its elongation at fracture is small, and its proof stress increases continuously as the temperature is lowered. A similar iron containing 0·15%P, water-quenched from 950° C, is brittle at a higher temperature. Its proof stress is not very different from that of the iron containing 0·05%P, but it becomes brittle at about — 80° C and its fracture stress then falls with falling temperature.[36] Again, the material breaks at — 196° C at a stress far below that at which it would be expected to yield. The brittle fractures are intercrystalline, and are fairly certainly due to a concentration of phosphorus at the grain boundary.[37]

**TABLE III**

**Fracture stress at — 196° C of 1%Ni alloys (A.C. 950° C, 2 h at 700° C, W.Q.)**

| O, % | 0·01 | 0·0019 | 0·0016 | 0·0006 |
|---|---|---|---|---|
| C, % | 0·0021 | 0·003 | 0·005 | 0·013 |
| Tons/in² | 12·3 | 26·4 | 42·2 | 49·0 |
| Fracture | Intergranular | | Cleavage | |

The two curves in the middle of the diagram are taken from the work of Wessel, and relate to mild steel.[38] In this case, embrittlement has been effected by the introduction of a deep sharp notch extending about one-third of the distance towards the centre of the specimen, and having a radius at the root of 0·0005 in. The unnotched specimen is ductile to below — 200° C. The notched specimen breaks in a brittle manner at — 150° C, and, again, the fracture stress decreases with decreasing temperature, and at — 196° C is well below the expected stress for general yield.

There is clearly something in common between these three types of brittleness. In all cases, the factor leading to the brittleness of the brittle member of the pair was present before the stress was applied. In the case of the notched specimen the brittleness is due to the stress concentration at the root of the notch, which restricts the volume in which plastic strain occurs more severely as the proof stress of the material becomes greater and consequently reduces the proportion of the strain energy that can be dissipated by plastic strain. In the two other cases there is no question of the presence of a gross stress concentration, and its place must be taken by some structural feature which serves the same purpose, either by supplying independently of the applied stress some of the energy needed to initiate a crack or by lowering the amount of energy that needs to be supplied.

It is in the nature of these structural features that the origin of toughness or brittleness in steels is to be found. They are obviously very different in different materials, but are probably present to some degree in all materials, since the unnotched mild steel, in which no recognized stress concentrator or embrittling agent was present, became brittle at about — 210° C, and at — 250° C broke at a stress well below the expected yield point. The stress at which they are capable of initiating fracture varies widely and the temperature at which brittleness appears depends upon this stress, and also upon the yield point of the material.

Finally, in Table III and Fig. 25 are given the fracture stresses at — 196° C of four quenched irons containing 1%Ni, and slightly differing amounts of carbon and oxygen.[39] As is to be expected, the carbon contents rise as the oxygen contents fall. All are brittle at — 196° C, but the fracture stresses rise consistently from one end of the series to the other, and whereas the two higher-carbon alloys have cleavage fractures, the two lower-carbon alloys have intercrystalline fractures. In spite of the difference of mode of fracture, the fracture stresses when plotted against the reciprocal of the carbon content lie on an unbroken line. This suggests, first, that the drop of strength is due to the oxygen in equilibrium with the carbon held in solution in the iron, and, secondly, that cleavage and intercrystalline fractures are not entirely distinct but are influenced by similar factors. It might not be fantastic to imagine that segregation of impurities to strained regions within the lattice is a cause of lowered cleavage strength, just as segregation to the grain boundary is thought to be a cause of lowered grain-boundary strength.

This is one experiment only, and may not be fully understood. But it does show how sensitive a property the fracture stress of a metal is, how easily it may be influenced, and how important it is to know the ways in which high fracture stress can consistently be obtained. For to raise the yield stress of a steel without ensuring that the fracture stress is sufficiently high is merely to produce a steel that is dangerous. Before the full strength of the iron crystal can be realized the origins of fracture must be found and their effectiveness reduced. It is not yet possible to say what and where they are, but it is surely not too bold to think that they will yield to sufficiently patient study. The persistent removal of minor imperfections has been the method of improving steels in the past and it appears that this also is the way for the future.

**Acknowledgments**

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**References**

1*a*. E. Goens and E. Schmid: *Naturwissensch.*, 1931, vol. 19, p. 520.
*b*. M. F. Markham: *Brit. J. Appl. Phys.*, 1957, vol. 8, Pt. s 56.
2. J. L. Snoek: *Physica*, 1939, vol. 6, p. 591; 1941, vol. 8, p. 711; 1942, vol. 9, p. 862.
3*a*. C. Wert and C. Zener: *Phys. Rev.*, 1949, vol. 76, p. 1169.
*b*. C. Wert: *J. Appl. Phys.*, 1949, vol. 20, p. 943; 1950, vol. 21, p. 1146.
*c*. C. Wert: *Phys. Rev.*, 1950, vol. 79, p. 601.
*d*. C. Wert: *Trans AIME*, 1950, vol. 188, p. 1242.
4*a*. N. P. Allen, W. P. Rees, B. E. Hopkins, and H. R. Tipler: *JISI*, 1953, vol. 174, p. 108.
*b*. W. P. Rees, B. E. Hopkins, and H. R. Tipler: *Ibid.*, 1951, vol. 169, p. 157.
5. W. Pitsch: *Acta Met.*, 1957, vol. 5, p. 175.
6. L. J. Dijkstra: *Philips Res. Rep.*, 1947, vol. 2, p. 357.
7*a*. G. I. Taylor and C. E. Elam: *Proc. Roy. Soc.*, 1926, vol. 112, ser. *A*, p. 337.
*b*. H. J. Gough: *Ibid.*, 1928, vol. 118, ser. *A*, p. 498.
8. C. A. Edwards and L. B. Pfeil: *JISI*, 1924, 109, p. 129; 1925, vol. 112, p. 79.
9*a*. C. S. Barrett, G. Ansel, and R. F. Mehl: *Trans ASM*, 1937, vol. 25, p. 702.
*b*. F. L. Vogel and R. M. Brick: *Trans AIME*, 1953, vol. 197, p. 700.
*c*. R. P. Steijn and R. M. Brick: *Trans ASM*, 1954, vol. 46, p. 1406.
*d*. R. Maddin and N. K. Chen: *Progress in Metal Physics*, 1954, vol. 5, p. 53; *Acta Met.*, 1954, vol. 2, p. 49.
*e*. *See also* A. Fahrenhorst and E. Schmid: *Z. Physik*, 1932, vol. 78, p. 383.
10. N. P. Allen, B. E. Hopkins, and J. E. McLennan: *Proc. Roy. Soc.*, 1956, vol. 234, ser. *A*, p. 221.
11. M. Hempel, A. Kochendörfer, and E. Hillnhagen: *Arch. Eisenhütt.*, 1957, vol. 28, p. 433.
12. B. Edmondson: Unpublished work.
13. E. A. Calnan and C. J. B. Clews: *Phil. Mag.*, 1951, vii, vol. 42, p. 616.

**References (continued)**

14. G. I. Taylor: *Proc. Roy. Soc.*, 1934, vol. 145, ser. *A*., p. 302.
15. J. Heslop and N. J. Petch: *Phil. Mag.*, 1956, viii, vol. 1, p. 866.
16*a*. H. W. Paxton and A. T. Churchman: *Acta Met.*, 1953, vol. 1, p. 473.
*b*. W. B. Biggs and P. L. Pratt: *Ibid.*, 1958, vol. 6, p. 694.
17*a*. C. A. Edwards, H. N. Jones, and B. Walters: *JISI*, 1939, vol. 139, p. 341P.
*b*. C. A. Edwards, D. L. Phillips, and H. N. Jones: *Ibid.*, 1940, vol. 142, p. 199P.
*c*. C. A. Edwards, D. L. Phillips, and Y. H. Liu: *Ibid.*, 1943, vol. 147, p. 145.
18*a*. M. Kuroda: *Sci. Papers Inst. Phys. Chem. Research, Tokyo*, 1938, vol. 34, p. 1528.
*b*. A. Nadai: "Plasticity": 1931. McGraw-Hill Book Co., New York and London.
19*a*. A. H. Cottrell: "Effect of Solute Atoms on the Behaviour of Dislocations": Phys. Soc. Report of Conference on Strength of Solids, Bristol 1947, p. 38.
*b*. A. H. Cottrell and A. T. Churchman: *JISI*, 1949, vol. 162, p. 271.
*c*. A. H. Cottrell and B. A. Bilby: *Proc. Phys. Soc.*, 1949, vol. 62, ser. *A*, p. 49.
20. R. H. Doremus: *Acta Met.*, 1958, vol. 6, 674.
21. A. H. Cottrell: *Trans AIME*, 1958, vol. 212, p. 192.
22. W. R. Hibbard and C. G. Dunn: *Acta Met.*, 1956, vol. 4, p. 306.
23. F. W. C. Boswell: *Met. Prog.*, 1957, vol. 72, p. 92.
24. C. C. Earley: Unpublished work.
25. L. B. Pfeil: *ISI Carnegie Schol. Mem.*, 1926, vol. 15, p. 319.
26. C. S. Barrett: "Cold Working of Metals," p. 65: 1948, Cleveland, ASM.
27. J. M. Berry: *Trans ASM*, 1957, vol. 51, Preprint No. 57.
28. National Physical Laboratory: Unpublished work.
29. J. J. Cox, G. T. Horne, and R. F. Mehl: *Trans ASM*, 1957, vol. 49, p. 118.
30. F. Osmond and C. Frémont: *Rev. Mét.*, 1905, vol. 2, p. 801.
31. J. F. Stead: *JISI*, 1898, vol. 53, p. 145; vol. 54, p. 137.
32*a*. J. R. Low: "Relations of Properties to Microstructure," p. 163, 1954, Cleveland, ASM.
J. R. Low: "Deformation and Flow of Solids," p. 60: IUTAM Colloquium, Madrid, 1955 (Springer, 1956).
*b*. N. F. Mott: *JISI*, 1956, vol. 183, p. 233.
*c*. W. S. Owen, B. L. Averbach, and M. Cohen: *Trans ASM*, 1958, vol. 50, p. 634.
*d*. N. J. Petch: *Progress in Metal Physics*, 1954, vol. 5, p. 1.
*e*. J. Heslop and N. J. Petch: *Phil. Mag.*, 1957, viii, vol. 2, p. 649.
*f*. A. N. Stroh: *Advances in Physics*, 1957, vol. 6, p. 418.
33. B. P. Haigh: *Trans Farad. Soc.*, 1928, vol. 24, p. 125.
34. G. W. Sears, A. Gatti, and R. L. Fullman: *Acta Met.*, 1954, vol. 2, p. 727.
35. A. A. Griffith: *Phil. Trans Roy. Soc.*, 1921, vol. 221, ser. *A*, p. 165.
36. B. E. Hopkins and H. R. Tipler: *JISI*, 1958, vol. 188, p. 218.
37. M. C. Inman and H. R. Tipler: *Acta Met.*, 1958, vol. 6, p. 73.
38. E. T. Wessel: *Proc. ASTM*, 1956, vol. 56, p. 540.
39. B. E. Hopkins and H. R. Tipler: *Rev. Mét.*, (in the press).

*Historical note no. 3 (second series)*

# The early ironworker and public opinion through the ages

*By H. R. Schubert*

EVER SINCE METALS were first used by prehistoric man, the smith was "endowed by popular feeling with magical potencies," on account of which he was dreaded by the Celtic peoples, just as the Druids and wise women were.[1] The awe-inspiring position which the smith held in the Saxon community of early England gave him a self-willed and independent character which occasionally brought him into conflict with the Christian Church. This is shown in the legend of St. Eegwin, Bishop of Worcester: when the Saint came to Alcester, in Warwickshire, around A.D. 700, the smiths of the town refused to listen to his preaching and attempted to drown his voice by the sound of their hammers and anvils.[2]

The smith could well afford to adopt an independent attitude since his work was very much in demand, not only for agricultural implements, but also for weapons. The activity of the Anglo-Saxon weapon-smith is illustrated in a manuscript written and illuminated in the reign of Edgar, King of Wessex in southern England (959–975) (*see* Figure). The smith standing to the left of the smithy-hearth has a hammer in his right hand for forging the iron held with tongs in a fire attended to by his assistant. The object they produced was apparently a spearhead. The spear with an iron head was the most common weapon of the Anglo-Saxons, and is shown carried by a troop of warriors equipped with such spears.

In the turbulent centuries of the Middle Ages the work of the skilled ironworker was greatly appreciated. There was hardly a castle without its own smithy. Further, large quantities of nails, horseshoes, missiles, and iron parts of siege machines (the forerunners of cannon) were made for the Royal Arsenal at the Tower of London. Smiths were directly attached to the army, in particular in the campaigns against the Scots, e.g. more than forty smiths in 1300–1302. They were recruited from various parts of the country, particularly from the Forest of Dean (16 smiths in 1319–20) and the Weald of Sussex and Kent.[3] They were also employed for the building of fortifications, e.g. in 1299 a master smith, Thomas de Furneys (by his name, perhaps, from Furness, in North Lancashire) with six smiths helped to erect a palisade at the castle of Dumfries in Scotland.[4] War-service was considered a great asset: on account of military service in an expedition against the Scots, William le Blomere, i.e. iron-smelter, from Tipton in Staffordshire was granted a complete pardon at Stafford in 1301, after he had mortally wounded another Tipton man with a poleaxe in a fight.[5] The appreciation of the medieval smith in general is expressed in the rhymed couplet of "The little smith of Nottingham who does the work that no man can."[6]

In the period of the Renaissance a mythological element was introduced into the description of the ironworker. Apart from his undoubted skill, it was on his sinister

appearance that a special stress was laid. The Italian architect A. A. Filarete in his Treatise on Architecture (written between 1460 and 1464) described furnacemen working in the Italian Alps near Pavia as strong fellows covered with soot, working in shirts or otherwise scantily clad, with their feet in wooden clogs, all in all resembling devils who torment the souls in Purgatory.[7] Nicolas Bourbon, whose father was the clerk (i.e. manager) of the ironworks at Vendeuvre in Champagne, in a Latin poem of 1517 compared the filler standing on the platform of the furnace-top to Charon, the ferryman of Hades.[8]

In periods of rapid industrial expansion the spoliation of the woods for the purpose of charcoal-burning was the main object of widespread grievances directed against ironworking. With the woods a good deal of wild life disappeared. In 1322 the keeper of Rothwell Park near Leeds, in Yorkshire, reported that the herons which had nested in the park and the sparrowhawks had been frightened away by the ring of the forge-hammers.[9] A complaint about disturbance of the peaceful countryside was raised 300 years later by Francis Somerford of Somerford Hall, north of Wolverhampton, in Staffordshire. He complained in 1623 that the beating of hammers in a forge erected only 80 yards from his mansion deprived him and his family " of their natural rest and sleep in the night." Embitterment caused by nervous strain may account for his slur on the ironworkers in general when he complained about " the ill neighbourhood of disordered and ill-disposed persons usually employed in such ironworks."[10]

It is only natural that some bad elements were amongst the workers in a period in which rapid expansion of the industry caused an extraordinary demand for labour such as in the boom period of the British iron industry in the early seventeenth century. When London iron merchants complained in 1631 about the bad quality of the bar iron produced at the ironworks of Richard Boyle, Earl of Cork, in South Ireland, Boyle replied that he had not much power over the workers who had become " in general so rich and proud, that I, my officers, and clerks may command them what we will, but they will do nothing but what they please."[11] Boyle, however, was sufficiently fair not to put all the blame on his workers. One of the reasons for their overbearing attitude, he admitted, was his " long absence from home." Another was, to use his own words, " my full and weekly paying my workmen."[12] This casts a peculiar light on the wages policy of employers in the early seventeenth-century iron industry. Irregular payment was by no means exceptional. In December 1609, for example, John Browne, founder at the Earl of Shrewsbury's ironworks near Sheffield, had to claim wages which were in arrears for more than half a year.[13]

The strength of the workers was always the pride of the iron industry.[14] It was in the interest of the employers to keep the men fit for their heavy work. There was a saying that the ironworkers " must be well lined with sack [cloth], strong beer and good victuals."[15] Drinks, such as ale and brandy, and food were lavished upon the workers on numerous occasions. This often led to drunkenness, which was a frequent source of complaint. The closing down of Ashburnham furnace in 1809 was ascribed to " the failure of the foundrymen, through intoxication, to mix chalk [as a flux] with the oar [ore], by reason of which it ceased to flow, and the blasting finally ended."[16]

At the awakening of Romanticism—the literary period in which the ' little man,' his character, life, and sufferings first were included in the frame of poetical motifs—a charming poem called *The Village Curate* set a monument to the ordinary village smith. It runs:

British Museum, Harleian Manuscript 603, fol. 6v.
**Anglo-Saxon smiths at work**

" The slaving smith,
That wipes his flowing brow so fast, his bread
Earns the bitter cost, expence of health.
In summer's hottest day, he feeds his forge
And stands expos'd to the distressful fire
That almost broils him dead. Yet what complaint
Makes he at fortune ? He is well content
To toil at his infernal work, and breathe
A torrid atmosphere, so he may earn
A scant subsistence in this pinching world."[17]

## References


1. R. J. FORBES: " Metallurgy in Antiquity," pp. 62 and 69: 1950, Leiden, E. J. Brill.
2. H. R. SCHUBERT: " History of the British Iron and Steel Industry," pp. 63–64, 1957, London, Routledge.
3. Ample evidence is supplied by the Royal Wardrobe Books, Public Record Office, Exchequer various accounts. See also E. STRAKER: " Wealden Iron," pp. 33–39: 1931, London, G. Bell and Sons.
4. British Museum, Additional Manuscript 35291, fol. 106.
5. Calendar of Inquisitions Miscellaneous (Chancery), vol. I, No. 2406, 1916, London.
6. H. R. SCHUBERT: Historical Note No. 12, *JISI*, 1949, vol. 162, p. 416.
7. W. VON OETTINGEN (ed.), *Auellenschriften für Kunstgeschichte, Neue Folge*, 1890, vol. 3, p. 473.
8. NICOLAI BORBONI: " Ferraria," published in the collection of his poems called " Nugae," 1533, Paris.
9. Public Records Office, Ministers Accounts (General series), bundle 1145, No. 21.
10. *Ibid.*, Chancery Proceedings James I, C 22, No. 51.
11. Chatsworth, Manuscript 78, pp. 373–374.
12. *Ibid.*, p. 193.
13. Shrewsbury Letters, City Library, Sheffield.
14. T. S. ASHTON: " Iron and Steel in the Industrial Revolution," p. 193: 1924, Manchester.
15. A. FELL: " The Early Iron Industry of Furness and District," p. 298, 1908, Ulverston.
16. E. STRAKER: *loc. cit.*, p. 369.
17. First published in 1788, reprinted by Straker, *loc. cit.*, pp. 304–305.

*PAPERS AND REPORTS ON RESEARCH AND PRACTICE*

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# Solidification characteristics of continuously cast low-carbon steels

*By M. Tenenbaum, C. F. Schrader, and L. Mair*

SYNOPSIS

The paper reviews the background leading to studies of the continuous casting process carried out by the Inland Steel Company and summarizes the results from these studies that have already been published. Typical cast structures are presented to show how solidification process has influenced operating problems as well as the properties and performance of the product. Detailed descriptions are given as well as longitudinal sections which illustrate selected features of the solidification process. A considerable part of the paper is devoted to structures obtained in continuously cast rimming steel sections. Typical segregation curves are also presented. On the basis of these data a mechanism is proposed for the solidification of continuously cast rimming steel. 1667

IN 1956, Inland Steel Company investigated the possibility of incorporating large-scale continuous casting facilities in an existing integrated plant producing carbon steels. The work was carried out with the co-operation of the Freyn Department of Koppers Company and Atlas Steels, Ltd. The primary effort in the Inland–Koppers–Atlas programme was directed toward continuously casting slab sections of low-carbon steel suitable for processing into hot- and cold-rolled sheet products. Before this co-operative programme, some exploratory studies had been carried out on large bloom sections continuously cast at the Mannesmann-Huttenwerke at Huckingen, Germany. Papers[1, 2, 3] dealing with engineering aspects, operating problems, production limitations, product properties, and performance in consumer plants have already been published.

In the earlier reports, reference has been made to aspects of solidification and structure which influenced machine design, plant operations, and product performance. It is the purpose of this paper to review the various types of cast structures that were observed, to indicate the operating factors contributing to major structural variations, and to propose mechanisms to explain the solidification characteristics of the continuously cast sections.

## CASTING OF LOW-CARBON STEELS

A major proportion of conventionally cast low-carbon hot- and cold-rolled sheet products are processed from rimming steel. Rimming steels occupy a prominent position in conventional casting because they can be produced without the use of any large deoxidizing additions; because reasonable surface quality may be obtained without resorting to costly preparation of the semi-finished section; and because a high ingot-to-product yield may be obtained by exercising moderate control during ingot teeming.

The most common deoxidizer added to low-carbon killed steels intended for sheet applications is aluminium. With careful control it is possible to develop cold-rolled aluminium-killed sheet products with improved resistance to ageing and, under certain conditions, with improved ductility. Usually, the presence of the excess aluminium content required to attain these improved properties leads to increased costs for surface preparation during processing. Furthermore, conventionally cast killed-steel ingots develop pronounced shrinkage cavities during solidification, and, as a result, often suffer a reduced ingot-to-product yield.

Many conventional processing and fabricating procedures for hot- and cold-rolled sheets are expressly designed to accommodate rimming steel. It can be recognized, however, that many of the characteristic advantages of rimming steel over killed steel would not necessarily apply when a continuous casting process is used. Specifically, the overall product yield would not necessarily be reduced by continuously casting killed steels instead of rimming steels. Accordingly, the studies described in this paper were designed to include a rather wide range of deoxidation levels. The structures of the sections produced in this programme, therefore, ranged from open porous castings typical of rimming steels to completely sound slabs that froze without visible gas evolution.

### Summary of earlier reports

Earlier reports have described the many operating problems encountered in casting low-carbon steel

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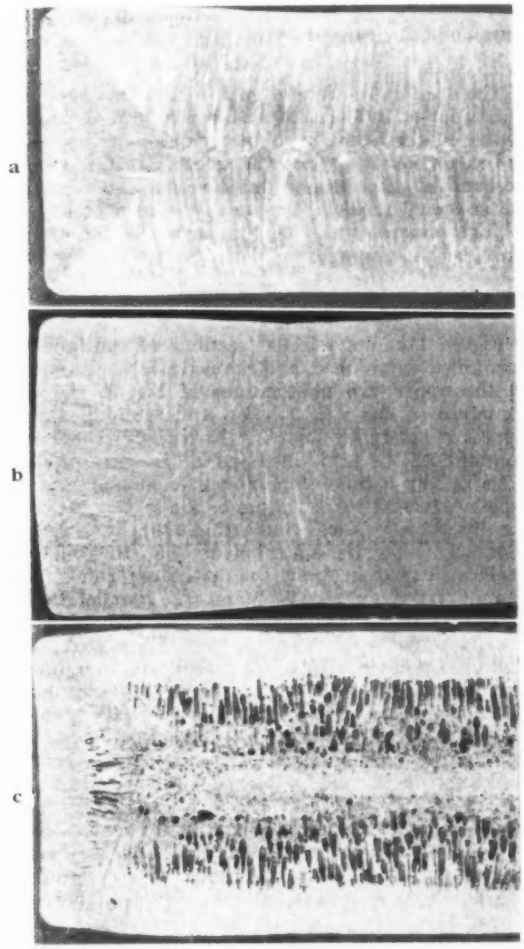
**TABLE I—Chemical analysis of continuously cast heats**

| Heat No. | Type of steel | Analysis, % | | | | | |
|---|---|---|---|---|---|---|---|
| | | C | Mn | P | S | Si | Other |
| 1085 | R | 0·09 | 0·33 | 0·008 | 0·025 | .... | .... |
| 3710 | R | 0·08 | 0·33 | 0·007 | 0·029 | .... | .... |
| 3733 | R | 0·08 | 0·23 | 0·010 | 0·035 | .... | .... |
| 3854 | R | 0·09 | 0·41 | 0·008 | 0·035 | .... | .... |
| 3879 | R | 0·07 | 0·28 | 0·016 | 0·034 | .... | .... |
| 3851 | R | 0·09 | 0·40 | 0·012 | 0·041 | .... | .... |
| 6977 | R | 0·07 | 0·25 | 0·009 | 0·022 | .... | .... |
| 7113 | R | 0·07 | 0·36 | 0·015 | 0·024 | .... | .... |
| 3794 | D | 0·08 | 0·37 | 0·011 | 0·039 | 0·07 | |
| 3795 | D | 0·07 | 0·43 | 0·012 | 0·029 | 0·01 | 0·033 Al |
| 3847 | D | 0·06 | 0·27 | 0·012 | 0·038 | 1·70 | |
| 3878 | D | 0·05 | 0·31 | 0·012 | 0·039 | 0·05 | 0·034 V |

R — Rimming D — Deoxidized

sections. It was brought out that problems were greatest when casting rimming steel. While the type of steel being cast influenced the rate at which a solidified slab could be withdrawn, the differences that were noted were relatively small. No attempt was made in the earlier reports to relate the actual cast structure to the casting rate.



(*a*) Killed steel, heat 3847; (*b*) Semi-killed steel, heat 3794; (*c*) Rimmed steel, heat 1085

**Fig. 1—Selected sections from continuously cast killed, semi-killed, and rimming steels**

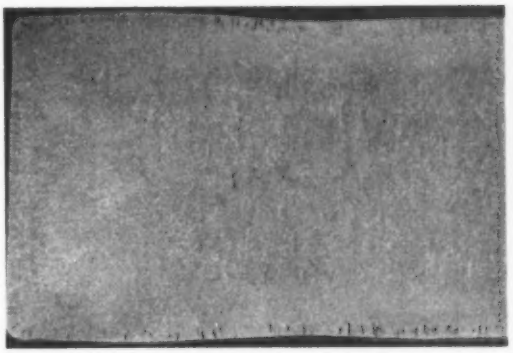


**Fig. 2—Structure of slightly under-deoxidized continuously cast semi-killed steel slab, heat 3878** × ½

Some photographs of selected sections were presented in earlier papers to illustrate the general nature of the structure obtained in continuously cast rimming steel. The work done with this type of steel indicated that difficulties could be expected with respect to the surface quality and internal soundness of the rolled product.

The casting of fully aluminium-killed steel was complicated by difficulties encountered in trying to obtain uniform metal flow into the casting machine and by intolerably poor surface quality. Between the deoxidation extremes, a substantial tonnage of silicon-deoxidized steel was cast with relative ease.

During processing of the three types of low-carbon steel it became obvious that if there were no shallow sub-surface blowholes, the cast structures responded very well to normal hot-working operations. The material in this programme was processed into hot and cold-rolled sheets, galvanized sheets, and tinplate. The rolled product was generally uniform in physical properties, and performed very well on selected commercial applications.

The steels cast in this programme were all made in electric furnaces. The product properties obtained reflected the residual element contents typical of the electric furnace process. When allowance was made for these residuals, the properties of the products were consistent with those expected in normal open-hearth material. Where quality aspects did not interfere, the expected high yield from liquid metal to final rolled product was realized. In many of the heats, however, losses because of poor rim steel structure or unacceptable killed-steel surface detracted substantially from the overall yield.

## CONTINUOUS CASTING OF LOW-CARBON STEELS

Part or all of 35 different heats were continuously cast during this investigation. The nominal heat sizes ranged from 6 to 30 tons. The cast product from each heat was sectioned and examined at a number of selected positions. Only 12 of the heats studied are specifically referred to in this paper. The ladle analyses of these are given in Table I.

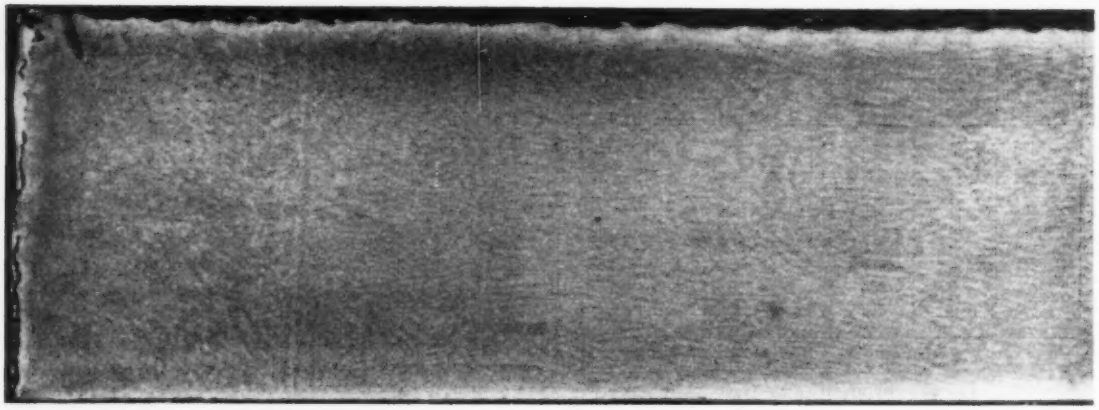


**Fig. 3—Structure near outer surface of continuously cast deoxidized steel slab, heat 3794. Etched Stead's reagent**

The equipment used to carry out these studies have been described in detail elsewhere[2, 5, 6] and need not be presented here. As in conventional casting, the steel was either deoxidized in the ladle or in both the ladle and the mould. The slab-section moulds used in this study were 24 in. × 6½ in. or 21 in. × 5½ in. In either case, the mould was 20 in. long.

The steel entering the mould formed a chilled outer skin next to the mould wall. The entire section was withdrawn steadily as metal continued to enter the mould. As the section descended below the mould it was comprised of an outer solid shell and a steadily shrinking liquid metal core. By the time the section was lowered 26 ft to the elevation of the cut-off torches, the core was completely solidified. With the withdrawal rates used in these studies, no difficulty was encountered freezing the complete section.

To assist in interpreting the structures observed in many of the cast sections it is desirable to indicate the magnitude of the freezing rates that were attained in this study. While it is not possible to give exact data regarding the speed of solidification, two reasonably well established base points are available. An earlier paper[2] brought out the fact that upon emerging from the water-cooled mould, the frozen outer shell is in the order of ½ in. thick. Using the classic Field equation,[4] it can be estimated that the solidification rate in the mould may be expressed by the formula:

$$D = 0{\cdot}8\,t^{1/2}$$

where $D$ = distance frozen, (in.), and $t$ = elapsed time, (min). While this expression is not necessarily exact, it indicates that the freezing rate in the water-cooled mould does not differ substantially from that observed in the early solidification stages of conventional ingots. The freezing rate in the mould has been verified by other investigators.[6]

To indicate the freezing rate in the spray zone it can be pointed out that even though withdrawal rates exceeded 60 in./min for limited periods, there were no instances where a section cut at the burning floor was not completely solid. Assuming that this is the limiting withdrawal speed and noting that the distance from the mould to the cut-off point was 26 ft, it may be estimated that the freezing rate formula in the spray zone is

$$D \geqq 1{\cdot}3\,t^{1/2}$$

The accelerated freezing rate in the spray zone was also reported by Speith.[6] While the form of the preceding equation may be subject to question, the values shown demonstrate that the freezing rate is greatly accelerated after the casting has emerged from the water-cooled mould. Presumably, the effective use of high-pressure water on the surface of the casting provided a medium for rapid continuous heat removal.

### Structures of three types of continuously cast low-carbon steels

Figure 1 shows selected sections of continuously cast killed, semi-killed, and rimmed steels. As shown in the upper two photographs of Fig. 1, the cast structure of low-carbon killed and semi-killed steel appeared relatively simple. Essentially, the structure of both the semi-killed and killed steel was characterized by a zone of prominent columnar crystals which extended to the centre of the section. Occasionally, there was some indication of slight tearing or separation along the central plane where the columnar grains growing in from the flat faces met. More commonly, there was some tearing parallel to the inner ends of the columnar grains. Unlike the situation in conventional casting, semi-killed steels were free of any obvious blowhole porosity. Accordingly, it was difficult to distinguish between the structure of continuously cast semi-killed and killed steels. The heavier deoxidation on semi-killed steel was necessary to ensure a usable product, since the first indication of porosity appeared in the form of skin holes (*see* Fig. 2). The greater deoxidation was not objectionable as it would be in conventional practice since it did not influence the yield of semi-finished product. The rimming steel section is characterized by a substantial outer zone apparently free of blowholes, a channel blowhole zone resembling a condition often encountered in the lower half of conventionally cast ingots, and a porous central core zone outlined by a continuous ring of rounded blowholes. Variations in operating and metallurgical

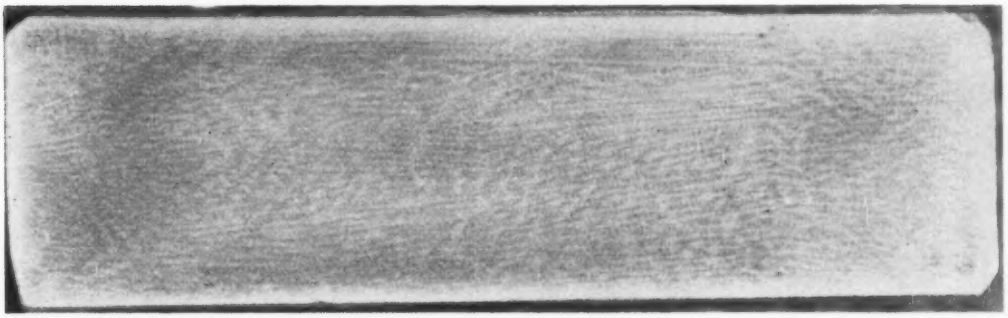


**Fig. 4—Structure near centre of continuously cast deoxidized steel slab, heat 3794. Etched Stead's reagent × 3**

conditions affected the relative size of these zones, but some indication of each could be recognized in all rimming steel sections examined.

The balance of this paper will be devoted primarily to discussion of the solidification characteristics of the continuously cast low-carbon steels. As indicated in the preceding section, the cast structures of killed and semi-killed steels were similar. These steels will therefore be considered under the major heading of deoxidized steels in the following section.

### Solidification characteristics of continuously cast deoxidized steels

As long as an adequate degree of deoxidation was maintained, there was little obvious influence of operating conditions on the overall structure of the deoxidized steels. The only consistent effect was a relation between corner tearing and withdrawal rate which was brought out in an earlier paper.[2] More detailed studies of selected heat-treated and etched* structures exposed a few subtle features of the primary freezing pattern.

Figure 3 shows the typical etched structure in from the surface of deoxidized steels. The photograph brings out the fine etched structure near the outer surface. Within the chill zone, is a pattern of fine columnar dendrites oriented roughly at right-angles to the outer surface. The dendrites gradually lose their directional tendencies at about ½ in. from the outer surface. Beyond this point, the freezing again becomes strongly directional. Figure 4 illustrates the manner in which the columnar dendrites extend to the centre of the cast section. In Fig. 4 the columnar dendrites meet along at a faintly defined plane toward the right end of the photograph. The section

* The sections were heated to 1 800° F, quenched in water, and tempered at 1 250° F, thus preserving the primary structure while erasing features associated with the austenite transformation.

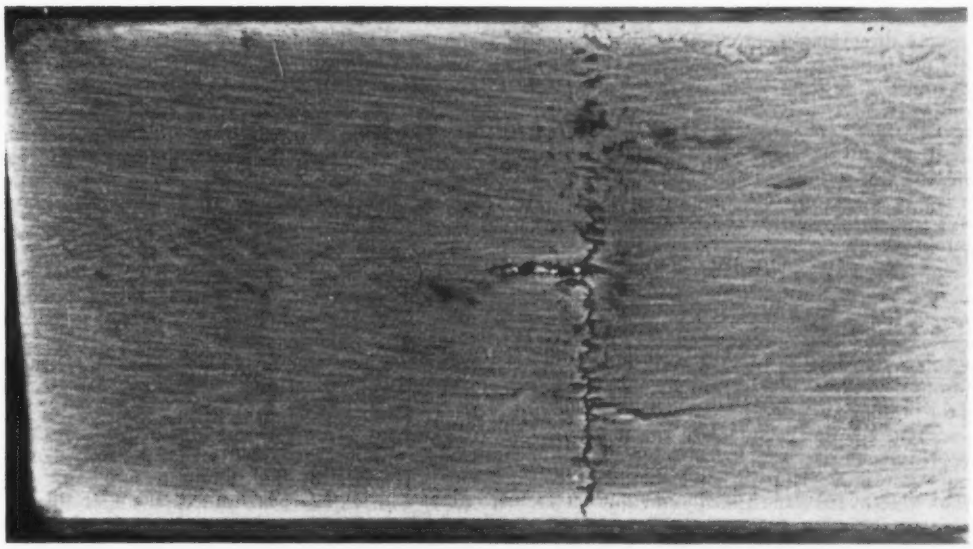


**Fig. 5—Structure near centre of continuously cast deoxidized-steel slab showing tearing at intersecting dendrite plane, heat 3795. Etched Stead's reagent × 3**

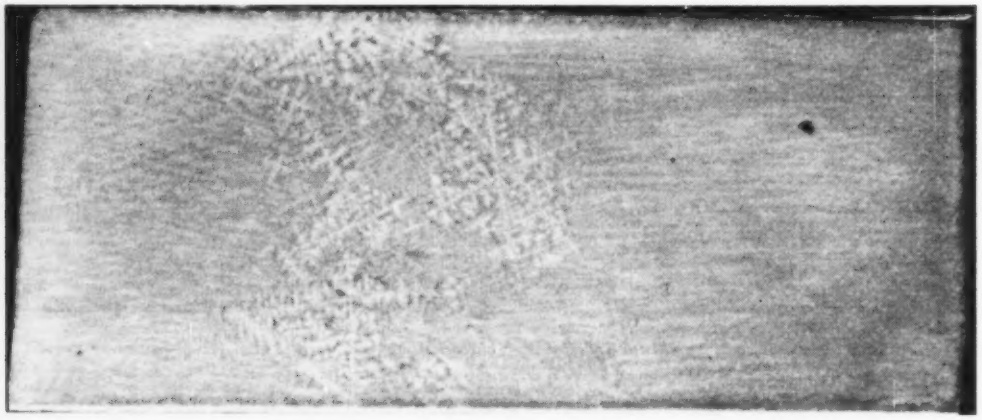


**Fig. 6—Random dendrite formation near centre of continuously cast high-silicon steel slab, heat 3847. Etched Stead's reagent** × 3

appears to be perfectly solid through the intersecting plane.

On numerous occasions, the plane at which the columnar dendrites met was not continuous. An example of a section which tore at the intersecting plane is given in Fig. 5. This photograph indicates the manner in which the dendrite pattern became more clearly defined at the centre of the section. Fortunately, the ragged central tear extended for only a short distance in the section. Figure 5 also shows a rather typical tear parallel to the colunar dendrites. Both types of tearing were encountered frequently in the killed-steel sections. The central tears did not have any serious effect on the soundness of the final product since they welded readily in later rolling operations.

The only deviation from the general patterns shown in Figs. 3–5 was noted in the structure of a low-carbon steel containing 1·70% Si. On this high-silicon steel, the strongly directional columnar freezing characteristics did not persist to the very centre of the sections. As shown in Fig. 6, the central structure of this steel was comprised of clearly defined randomly oriented dendrite patterns. Tentatively, this situation is associated with a difference in the heat transfer characteristics of the high-silicon steel.

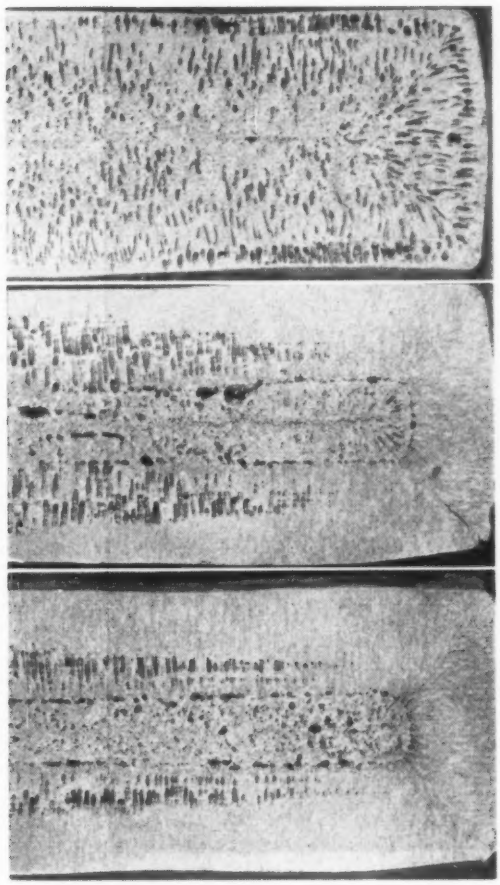
#### Solidification characteristics of continuously cast rimming steel

The rim steel heats produced for this programme were all tapped at a low bath carbon content. Lance oxygen and iron ore were used for carbon removal. Because of difficulties encountered in co-ordinating steel refining and casting plant activities, it was not uncommon to hold heats in the furnace at a low carbon for prolonged periods.

The standard ladle deoxidation for rimming steel to be continuously cast was 0·6 lb Al per net ton of liquid metal. In experiments directed toward evaluating the effect of ladle deoxidation on rim thickness and the control of metal action in the mould, the ladle aluminium was varied from 0·5 to 0·8 lb/net ton. This level of ladle deoxidation is recognized to be considerably higher than that which is associated with conventional casting of rimming steel. Three factors contributed to this higher aluminium requirement. First, the metal was often held in the furnace for a prolonged period at low carbon, and this led to increased oxidation. Second, the slag was removed from the furnace before tapping; and for a brief period, the bare metal surface was exposed to the oxidizing furnace atmosphere, a condition which also was conducive to a higher oxygen level in the steel. Finally, to ensure positive control of metal action in the continuous casting mould, it was necessary that gas evolution be significantly less vigorous than is desired in the conventional casting of rimming steel.

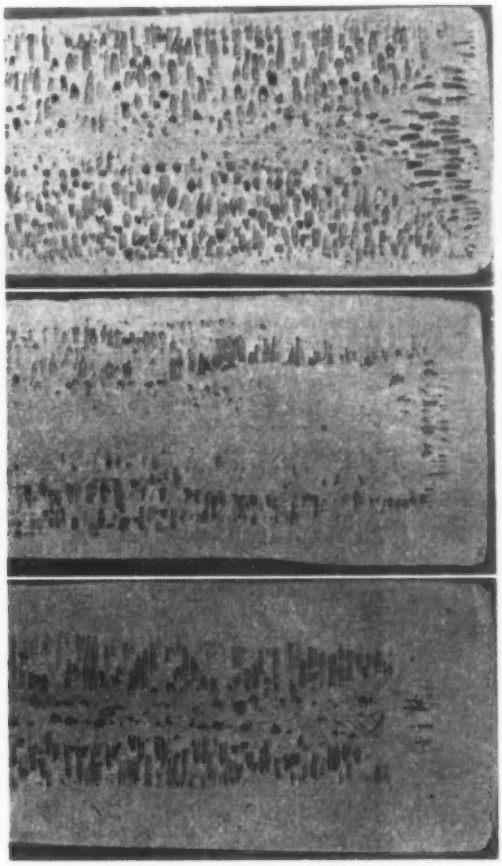
As in conventional casting, some minor adjustment in oxidation was necessary after the metal left the ladle. In this investigation, the final deoxidation was accomplished as the metal entered the mould. This was done to permit direct observation of the influence of changes in oxidation on gas evolution and cast structure. In routine operations it is likely that this final deoxidation adjustment would be carried out before the time the metal entered the mould. This final deoxidation was most critical, since it was necessary to develop sufficient action to secure an adequate rim zone while still avoiding the excessive boiling and foaming that result characteristically when oxidized metal contacts a cold mould surface. In this respect, the traditional system of feeding aluminium pellets into the mould was not acceptable for the continuous casting process. To obtain a uniform and closely controlled deoxidation level, an aluminium wire feeder was developed which was capable of feeding the deoxidizer into the stream entering the mould at a carefully measured and uniform rate.

Probably the most critical aspect of rimming-steel structure concerned the thickness of the solid outer rim. Where this rim was too shallow, the blowholes were exposed on reheating and rolling. This, in turn, resulted in a severely checked and torn surface condition in the rolled product. For any given operating conditions, the thickness of the solid outer rim depended on the vigour of gas evolution in the



| *Slab No.* | *Al addition, lb/ton* |
|---|---|
| (*a*) 5B | 0·55 |
| (*b*) 3T | 0·39 |
| (*c*) 2B | 0·35 |

**Fig. 7—Effect of increasing mould aluminium additions on structure of continuously cast rimming-steel slabs, heat 3733** × ¼



| *Slab No.* | *Casting rate, in./min* |
|---|---|
| (*a*) 5B | 40 |
| (*b*) 3B | 34 |
| (*c*) 1B | 26 |

**Fig. 8—Structure of slabs continuously cast at three different withdrawal rates, heat 3879** × ¼

mould. As in conventional casting, the more highly deoxidized the steel, the closer the blowholes approached the outer surface. This situation is illustrated in Fig. 7, which presents sections from a series of slabs from a single heat. The slab represented by the sections in Fig. 7 were deoxidized with varying amounts of aluminium. When the mould aluminium addition was decreased from 0·55 to 0·35 lb/ton, the solid outer rim increased from practically nothing to a thickness of about 2½ in. Further reductions in mould deoxidation were not feasible since the resultant heavy gas evolution would not permit accurate regulation of the metal level in the casting mould. The middle photograph also gives a good illustration of a tear in the corner of a slab section.

The control of gas action in the mould and the regulation of the liquid metal level were complicated when attempting to increase the casting speed. At low speeds, it was possible to alter the mould aluminium addition and thus vary the internal structure over a relatively wide range. As the speeds approached and exceeded 35 in./min, the action of the liquid steel became erratic and frequent surges in gas evolution introduced a definite hazard of the metal level rising out of the mould. Some of these surges could be accommodated by rapid adjustment of slab withdrawal rate, but this erratic type of operation did not allow any latitude in the control of internal structure or surface quality. A selected example indicating the effect of casting speed is shown in Fig. 8. In this heat, an attempt was made to maintain a uniform and controlled rate of gas evolution. It can be recognized from Fig. 8 that the conditions associated with the higher withdrawal speeds gave a very thin-skinned structure in the cast slab.

An illustration of the improved structure attainable with low withdrawal speeds is given in Fig. 9. The section shown in Fig. 9 is from a 13-in. square bloom withdrawn at only 12 in./min. At this slow withdrawal speed, it was possible to develop a substantial depth of sound outer rim despite an extremely mild action in the mould. It can be surmised that the low rate of gas evolution and the resultant moderate

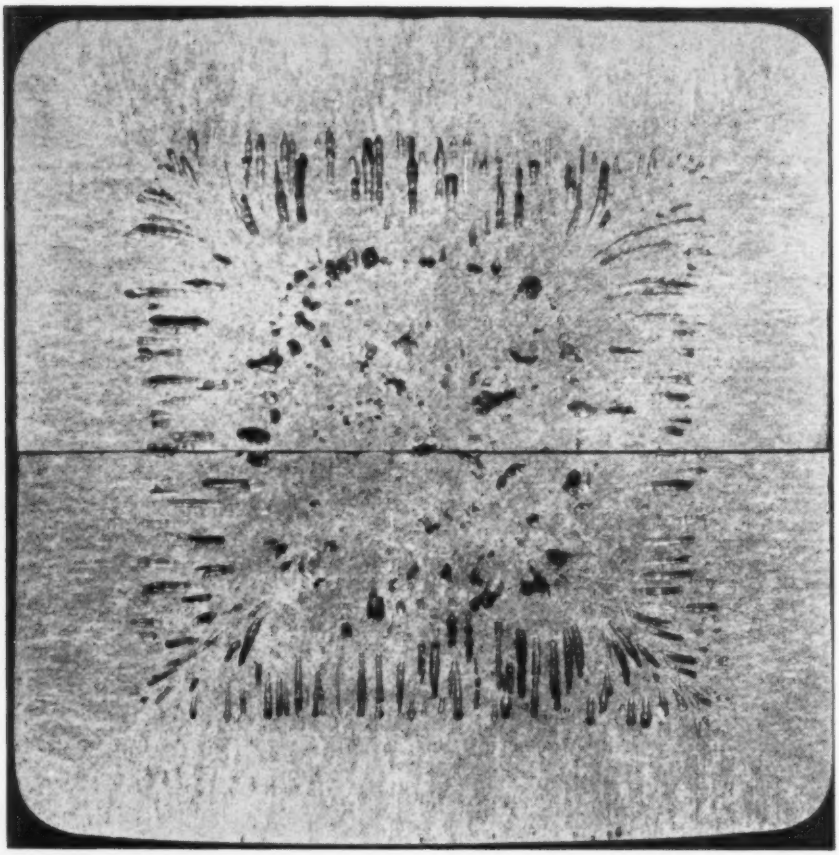


**Fig. 9—Structure of low-carbon rimming steel bloom section continuously cast at a withdrawal rate of 12 in./min, heat 6977. Mould action very mild** × ½

liquid metal action that were developed in casting the large bloom section of Fig. 9 would have contributed to an extremely thin-skinned structure in slabs cast at 35 in./min.

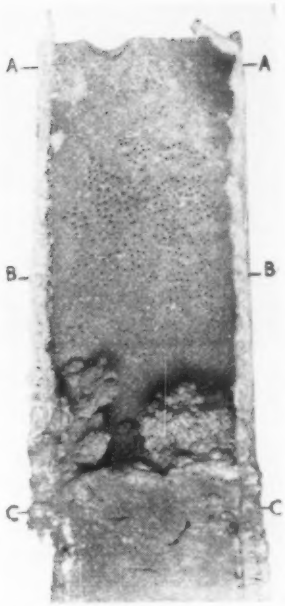
An interesting example of thin-skinned structure in a slab that was withdrawn at an excessively high rate is shown in Fig. 10. The slab sectioned for Fig. 10 was withdrawn at a speed close to 60 in./min. This speed was so great that it did not permit formation of an adequate shell thickness, and, as a result, the liquid metal broke through the surface of the casting below the level of the water-cooled mould. When the metal ran out, the level within the shell dropped, exposing the solidifying faces shown in Fig. 10.

In an effort to increase the thickness of solid skin in the slab sectioned for Fig. 10, the mould deoxidation had been reduced to the point where the metal action was extremely erratic and difficult to control. Despite this vigorous action, blowholes were entrapped close to the surface. The beginnings of the channel blowhole structure can be seen in Fig. 10 along with the steadily increasing depth of the shell below the surface. For some unexplained reason, one corner solidified more slowly than did the balance of the section. The break-out occurred through this narrow section.

*Rim zone*

Figure 11 shows the detailed structure of a section taken from the rim zone of the middle slab of Fig. 7. This slab was withdrawn at a rate of about 37 in./min. The structure is characterized by almost complete absence of any visible dendritic pattern. Very light rim-hole contour patterns can be recognized approaching the channel blowholes and along the segregate streaks in the outer solid rim. The rim-hole contours indicate a sequence of successive gas bubble formations and evolution as the channel-hole zone is approached. As freezing progressed, the gas bubbles no longer were completely evolved and the continuous channel structure developed.

In contrast to the preceding photograph, the section from which Fig. 12 was taken was withdrawn at about 12 in./min. This section shows definite evidence of dendritic structure especially at the earlier stages of freezing. In this section, the rim-hole contours progress inward for a considerable distance and are easily visible. In certain areas, these rim holes have been completely closed leaving a solid rim zone or a small blowhole remnant. An excellent




10a

10b

Fig. 10—Break-out sections (*b*) showing gas entrapment at start of freezing of outer shell, heat 3854: Sections as indicated in (*a*) × ¼

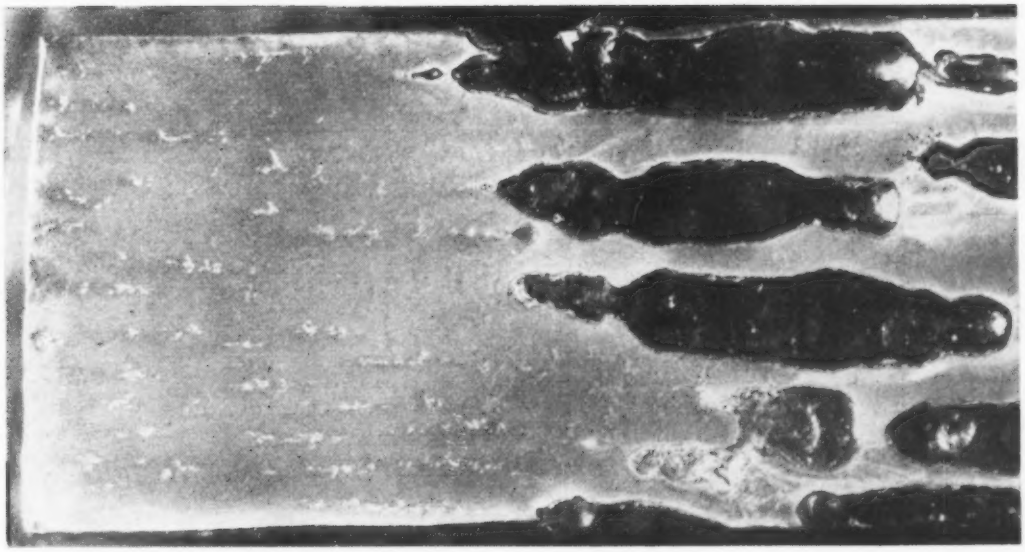


Fig. 11—Structure of rim zone of continuously cast rimming-steel slab, heat 3733. Etched Stead's reagent, withdrawal rate 37 in./min. Note contours approaching channel blowholes × 3

**Fig. 12—Structure of rim zone of continuously cast rimming-steel bloom, heat 6977. Etched Stead's reagent, withdrawal rate 12 in./min. Mould action very mild** × 3

example of a partly refilled channel blowhole is shown in the lower part of the section. As in the preceding section, the rim-hole contours lead into a channel-type blowhole with the cross-section broadening as the distance from the outer surface increases.

Figure 13 was prepared to indicate corner solidification characteristics for sections 5B and 3T of Fig. 7, both of which showed evidence of tearing. The slab of Fig. 13 was heavily deoxidized and, as a result, was thin-skinned with short blowholes forming

(*a*) Slab 5B, over-deoxidized; (*b*) Slab 3T

**Fig. 13—Structure of corner sections of two continuously cast rimming-steel slabs from heat 3733. Etched Stead's reagent** × 2

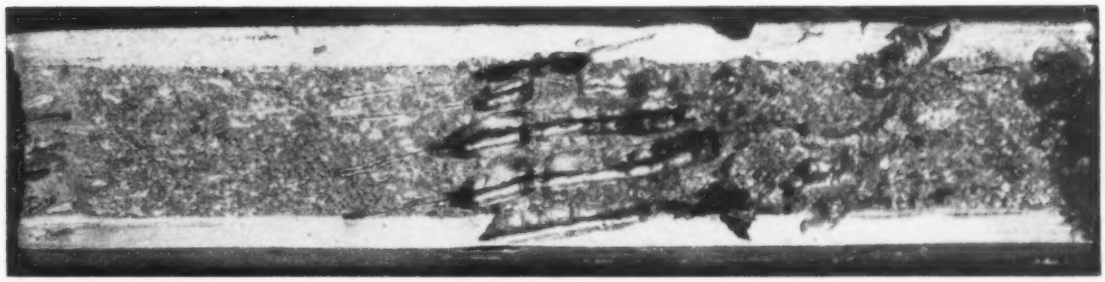


**Fig. 14—Broken section of continuously cast rimmed-steel bloom showing chill effect near outer surface, heat 7113** × 3

very early in the freezing process. The corner tear does not appear to be related to the early blowhole formation. The illustration is interesting since it can be recognized that there were alternate periods of gas entrapment followed by gas evolution. The entrapment is indicated by the blowhole pattern, while the evolution is often marked by faint outlines of successive contours. Toward the centre of the section are several smaller blowholes with contours extending out obliquely from the freezing direction. Presumably, these blowholes represent the remainder of what originally was a series of larger bubbles.

Figure 13*b* was taken from the torn corner of the middle section (3T) of Fig. 7. This section was intentionally less highly deoxidized than was the preceding slab. The traces of a series of blowholes are evident in Fig. 13, all pointing to the diagonal of the section and curving in towards the centre of the slab. The photograph brings out a faint pattern of radial streaks which parallel the outer surface. Again there is no evidence of any relationship between gas formation and the corner tear. Presumably the tear occurred after the corner had solidified and is a result of stresses imparted because of shrinkage during the cooling of the outer shell. It is interesting to note that the radial streaks are curved so as to remain roughly at right-angles to the segregate streaks. Approaching the inner blowhole zone a system of very faint blowhole contours could be detected. Only a vague suggestion of this contour pattern can be recognized in the photograph. These contours are aligned along the radial streaks. While there is no positive evidence to support such a deduction, it is proposed that the radial streaks represent successive positions of the freezing interface where there were surges in the liquid metal in the moulds.

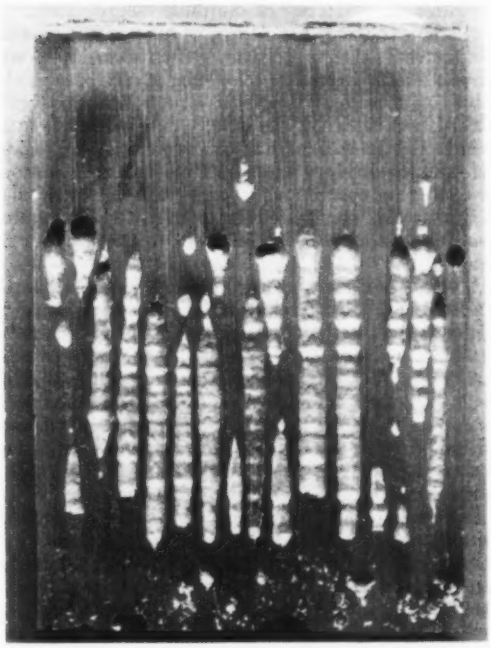


**Fig. 15—Successive expansion and contraction pattern in continuously cast low-carbon rimming steel, heat 3851 (section shown standing on side).** × 1

The large tear of Fig. 13*b* is located almost exactly on the diagonal. The parting cavity has a typical ragged appearance that would be expected of a high-temperature tear. From its location, it is assumed that the tearing took place along the intersection of the faces solidifying in from two adjacent right-angle surfaces. Presumably, this intersection is a zone of weakness in the overall cast structure. There is no obvious connection between the corner tears and the rupture within the casting.

Many of the illustrations that have been presented indicate that gas formation takes place during early stages of solidification despite the extremely rapid freezing rate. It has often been proposed that in the initial stages of solidification of conventional rimming steel ingots, the chill effect actually suppresses gas formation. It is interesting to recognize, therefore, that although the oxidation levels in these continuously cast steels were low, and despite the fact that conditions were conducive to drastic initial chilling, it was not possible to suppress skin-hole formation. A good example of skin-hole formation in the chill zone is shown in Fig. 14. This figure was prepared simply by breaking a small section to expose the internal structure. A system of skin holes is obvious in the severely chilled zone next to the surface. This outer chill is followed by a zone relatively free of any blowholes. Further into the section, the blowholes reappear in the form of thin continuous cavities which eventually widen out into rather typical channel blowholes. Skin holes of the type shown in Fig. 14 proved very troublesome throughout

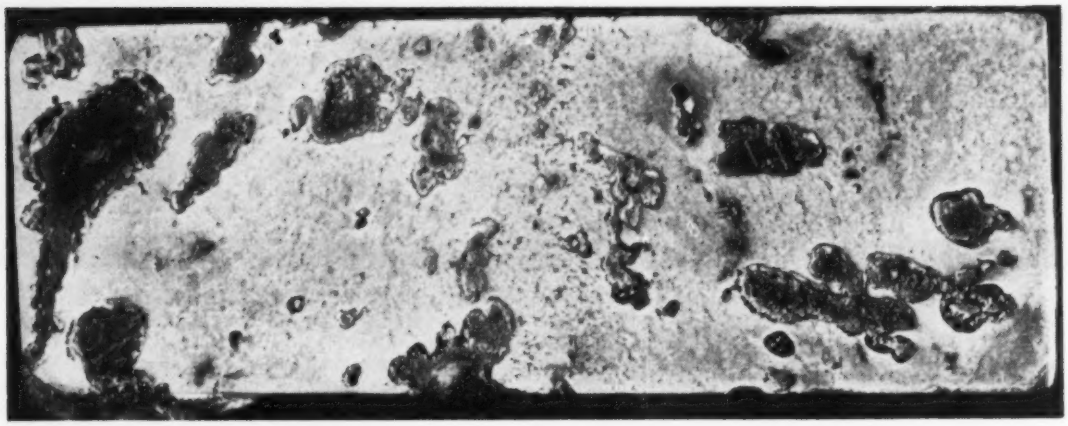


**Fig. 16—Core zone structure in continuously cast rimming steel, heat 3710. Etched Stead's reagent** × 3

this investigation, because of their consistent association with seam- and sliver-type defects in the rolled product.

The channel blowhole structure developed in the continuously cast slabs were similar to those encountered in the rim zone near the bottom of conventionally cast ingots. The channels are characterized by periodic contraction and expansion of the cavities reflecting a sequence involving alternate entrapment and evolution of gases. Similar contraction patterns can be recognized in adjacent channels indicating that gases were evolved simultaneously over a considerable length of the solidifying face. This condition can be recognized vaguely in Figs. 1, 7, 8, 9, and 14. Figure 15 shows the alternating contraction and expansion pattern clearly. The channels always extended a considerable depth into the slab frequently terminating at a rather clearly outlined 'core zone'.

There was no direct effort to cap the casting as is done in conventional practice. It is presumed that by the time the outer shell had frozen in as far as the outer edge of the core zone, the accumulated head of liquid metal was sufficient to suppress the carbon–oxygen reaction to the extent that gases no longer freely evolved. It might be expected, therefore, that the thickness of the rim zone would be related to the slab withdrawal rate. No evidence was found to indicate that such a relationship actually existed.

*Core zone*

Figure 16 shows a typical core zone pattern for continuously cast rimming steel. The deep etched structure is characterized by scattered porosity and complete absence of any columnar dendrite pattern. The very centre of the section displays a more marked segregate pattern than could be seen in any other zone in the rimming steel section. For the most part, this segregate was concentrated within a relatively narrow band at the centre of the section.

**Segregation curves**

Several slabs from each heat made in this investigation were sampled extensively to explore the segregation characteristics of the continuously cast slabs. An attempt was made to sample all slabs at the same relative positions. For a given type of steel, the segregation curves for each section sampled were remarkably consistent. Accordingly, all segregation data for steels of each type were combined in the curves of Fig. 17. No attempt is made to distinguish between semi-killed and killed steels since the segregation curves for the two types were practically identical. While complete analysis was obtained on each sample, only the data for sulphur and carbon contents are plotted in Fig. 17.

On the rimming-steel slabs, there was an easily recognizable pattern of sulphur segregation. As might be expected, the degree of segregation was not pronounced relative to the values associated with conventional ingots. The negatively segregated outer rim extended about half-way in from the surface towards the centre of the slab. This feature is interesting since it means that the negatively segregated rim comprised more than half of the cross-sectional area of the continuously cast rimming-steel slab. This may be contrasted to conventional rimming-steel ingots where about one-third of the area is rim zone.

A curve is also shown in Fig. 17 to illustrate the extent of carbon segregation in rimming steels. In this case the amount of segregation is significant but relatively small.

As brought out in earlier discussions, all deoxidized steels, whether semi-killed or killed, were cast so as to avoid blowhole formation anywhere in the cast section. It is interesting to recognize that there was no significant segregation of carbon and sulphur in these continuously cast deoxidized steels.

Segregation data were also obtained for manganese and phosphorus. Even in rimming steels, these elements did not segregate sufficiently to permit detection by the sampling and analysis procedures used in this study.

**Solidification of continuously cast low-carbon steel**

To provide a basis for speculating on the mechanics of freezing in continuously cast rimming or deoxidized

low carbon steel sections, the following listing has been made of selected solidification characteristics that have been recognized in the structure of conventionally cast ingots.

1. Columnar freezing occurs where definite temperature gradients exist at the solidifying interface.[8] The continuous casting process provides an opportunity for maintaining high-temperature gradients since the mould wall does not heat up as in conventional practice. Even more important, a large fraction of the heat is removed by water being sprayed on to the casting surface[2] thus avoiding air gaps which could retard the heat-transfer process.

2. In the absence of gas formation and evolution, columnar freezing can be recognized by a characteristic system of dendrites which roughly indicates the direction of heat transfer. Microsegregation phenomena make it possible to recognize columnar dendrites in the solidified section.

3. Most of the gas formed during solidification of low-carbon rimming steels results from reaction between dissolved carbon and oxygen.

4. Because of the rapid initial rate of heat removal, the outer skin of an ingot freezes very rapidly. In continuous casting, this chilling action would be favoured by the low temperature and high conductivity of the mould wall. The rapid freezing rate could conceivably inhibit the gas-forming reactions. Under certain circumstances, rapid freezing would make it difficult for gas that has formed to be evolved.

5. Vigorous gas formation and evolution tends to eliminate the micro-segregation that makes it possible to recognize dendritic structure in solidified castings.[9]

6. When all the gas formed during freezing is evolved, the solidified section is free of entrapped blowholes.

7. When substantial quantities of gas are formed and entrapped during rimming, the outer rim is characterized by a system of adjacent channel blowholes aligned in the general direction of heat transfer.

8. Where a small fraction of the gas formed during rimming is entrapped, the diameter of the channel blowholes is considerably reduced.

9. Gas formation is inhibited by high ferrostatic head and low dissolved-oxygen content.

10. Under certain conditions, gas bubbles are alternately formed and evolved during rimming, and this phenomenon can often be recognized by a pattern of contours outlining the perimeters of the original bubbles.

11. Wholesale gas evolution along a freezing interface is accompanied by definite negative macrosegregation in the solidified rim and positive segregation in the remaining liquid metal. The pattern of sulphur distribution across the solidified section provides a reliable index of the tendency for such segregation to occur.

The observations that were made in examining the various sections coupled with characteristics that have been recognized in the structure of conventional ingots provide a basis for speculating into the mechanics of solidification of continuously cast low-carbon steel. This discussion is confined to deoxidized steels which freeze with no visible entrapped gas in the final section and to continuously cast rimming steel. The mechanics of solidification not only differs with steel composition and oxidation level, but also must vary with the dimensions of the section being cast and the construction of the casting machine. The nature of this study necessarily limits consideration to the products cast in a single machine.

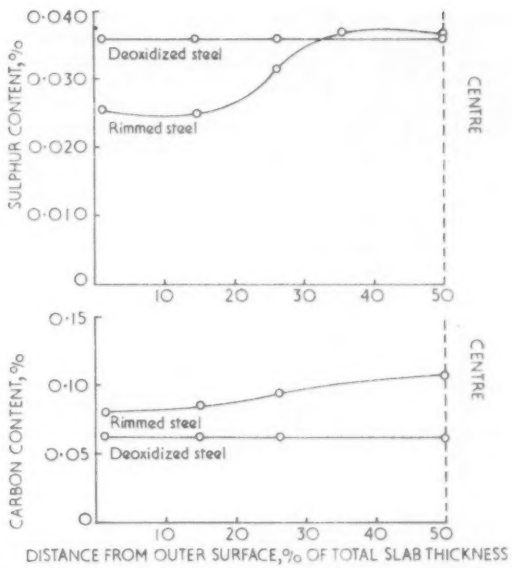



**Fig. 17—Averaged sulphur and carbon segregation curve for continuously cast rimming and deoxidized steel slabs. Centre-line represents centre of slab**

### Solidification of continuously cast deoxidized low-carbon steel slabs

The observations made in this study indicate a simple mechanism for solidification of deoxidized steels. On such steels, an initial chill was formed at the surface which resembled the outer skin of a conventionally cast ingot. As the section was lowered, the rate of heat removal was retarded and a relatively uniform freezing interface was developed at which a definite temperature gradient existed. Under these conditions, highly directional columnar freezing started. While the section was being withdrawn in the water-cooled mould, the thickness of the solidified shell increased and some shrinkage occurred. As in conventional casting, this initial shrinkage caused the formation of an insulating gap between the frozen shell and the inner surface of the water-cooled mould. The rate of heat transfer was retarded and the temperature gradient at the freezing interface was reduced. By the time the section reached the bottom of the water-cooled mould, the columnar freezing characteristics became less pronounced.

Upon emerging from the mould, the outer surface was cooled by the heavy water-spray system, and the sharp temperature gradients reappeared. Because of the continuous effective heat removal accomplished in the water-spray zone, the temperature gradient at the freezing interface persisted until the section was completely solid. Under these conditions, columnar dendrites extended inward until they met the surface freezing in from opposite or adjacent

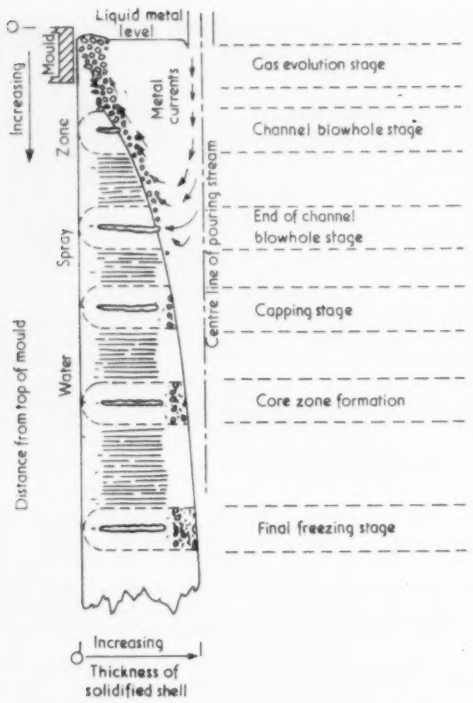



**Fig. 18—Schematic sketch showing stages in the solidification of a continuously cast slab**

walls. Occasionally, where there were discontinuities in the casting or where the steel grade being cast had lower heat conductivity, some evidence of random dendritic freezing was observed at the centre of the section.

The solidification of continuously cast rimming steel, on the other hand, was complicated by the interrelated effects of a metal stream being fed into the water-cooled mould at a constant rate and a steadily increasing head of metal over any given section.

When the continuous casting process was operating steadily and the solidified bars were being cut off and discharged from the strand, a situation was reached where a deep cone of liquid metal was contained in a solid shell. This shell varied in thickness from practically zero at the point of metal entry to a completely solid section at some point far below the mould level. A schematic illustration of the condition that exists when this 'steady-state' operation was attained in the casting of rimming steel is given in Fig. 18.

Unquestionably, the outer skin of the upper metal layer chilled very rapidly on contacting the water-cooled mould wall. The extent of this chill zone apparently varied within a heat and from heat to heat. On some sections observed, the depth of chill was not perceptibly greater than that associated with rimming steel poured into conventional moulds whereas, on others, there was evidence of drastic initial cooling rates. The rapid cooling during the early stages of freezing was not sufficient to suppress gas formation in the skin. On the other hand, the freezing rate was fast enough to entrap some of the gas formed at this stage, and this situation often resulted in the presence of shallow blow-holes very close to the outer surface of the casting.

As the casting was lowered in the water-cooled mould, the chilled shell contracted. Presumably this contraction produced a gap between the casting and the mould wall which retarded the rate of heat transfer. This interference with the heat transfer process slowed down the rate of solidification in the mould to the point where it was not markedly different from that associated with the early stages of freezing in conventionally teemed ingots.

As the freezing rate slowed, the gas which formed along the solidifying interface could be more readily evolved. To encourage such gas evolution, it was necessary to regulate the oxidation level carefully. The qualitative aspects of this control were apparently similar to those involved in regulating gas evolution in conventional rimming-steel ingot practice. It is assumed that the continuous stream entering the mould caused currents of liquid metal to circulate along the freezing interface which helped release the gases formed just inside the skin. The presence of such liquid metal currents in continuously cast sections was recognized by Krainer and Tarmann.[10]

For the most part, the gas bubbles that formed during the early stages of freezing were dislodged by the combined effects of the rising metal currents and the relatively vigorous gas-forming reaction. Where the resultant rim was solid, it was quite difficult to detect any blowhole contour pattern in the outer rim section of the solid casting. Under conditions of less vigorous metal action, it was possible to recognize characteristic contours marking the position where the gases which formed during freezing were evolved and the original blowhole refilled with liquid metal.

It was estimated that about $\frac{1}{2}$ in. of outer shell was formed while the casting was in the water-cooled mould. With reasonable withdrawal rates it was possible to control metal deoxidation so as to obtain a sound structure in this outer shell. On rimming steels there was no obvious structural change that could be related to the time at which the section entered the water-spray zone. In many sections, a series of long channel holes appeared at the distance from the outer surface which conceivably could have corresponded to the time of entry into the spray zone, but the exact position could be varied readily by changing the rate of deoxidation. It was concluded that while the position of these channel blowholes may have been influenced by the change in heat transfer rate, the effect was masked by features such as oxidation level and withdrawal rate.

As the section was withdrawn further, it was covered by a continuously increasing head of metal. The rising head of metal tended to suppress gas formation and evolution. As in conventional practice, the less vigorous gas evolution resulted in bubble entrapment with the formation of long continuous-channel blowholes. Simultaneous contraction and expansion of these channels along planes parallel to the surface indicate a pattern of bubble formation

and partial release similar to that encountered in the lower half of rimming ingots. In the continuously cast sections, the pattern may have been influenced by changes in the velocity of the stream entering the mould and the manner in which metal currents were distributed throughout the deep liquid-metal cone.

The rim patterns encountered in this study varied considerably, the structure ranging from several inches of solid metal to a system of continuous channels from the extreme surface to the central core. In all cases, however, this rim was negatively segregated throughout the outer solid rim and the channel blowhole zone. While the degree of segregation was not as great as in conventional ingot practice, the relative proportion of the volume of the final casting which was negatively segregated was considerably greater. In any case, the extensive segregation that occurred indicates that substantial volumes of gas were evolved during freezing of the rim zone.

With further withdrawal of the section (and as the solidified outer shell occupied an increasingly large fraction of the cross-section), the distance from the upper surface became so great that the freezing interface was no longer exposed to currents generated by the entering stream. Because of the high head of metal, moreoever, the gas-forming reaction was suppressed to the point where bubbles could not be evolved spontaneously. At this point, a structural condition was observed in the continuously cast section which resembled that obtained when conventional rimming ingots are capped. The capping position was usually marked by a pattern of rounded blow-holes outlining a central porous core zone. As in conventional practice, there was a definite change from negative to positive segregation at the position of the blowholes outlining the core zone.

Gas formation within the core zone appeared erratic, presumably being influenced by local concentrations of the reacting elements, the presence of scattered nuclei, and the exact manner of operation of the casting machine. The various blowhole systems that encircled the core zone unquestionably interfered with the heat transfer process, and, as a result, there was no evidence of any columnar freezing at the centre of the section. Instead, the structure was comprised of a well defined globular segregate pattern which clearly marked the position of the last metal to freeze.

## SUMMARY

This paper has reviewed various structures observed in continuously cast low-carbon steels. In so far as structure was concerned, the steels were divided into two basic types, deoxidized steels and rimming steels.

It was brought out that the structure of deoxidized steels was fairly simple. The solidification of these steels was characterized by columnar freezing from the outer surface to the centre of the section. The columnar freezing reflected the definitely directional pattern of heat removal which was characteristic of the continuous casting process. A slight interruption in this freezing pattern could be recognized at the position where the section passed from the water-cooled mould into the water-spray zone. There was very little macroscopic segregation observed in the deoxidized steels.

The solidification pattern on rimming steels was more complex. Gas-forming reactions occurred in a manner similar to that encountered in conventional ingot practice. The effectiveness of removal of such gases from the casting was regulated primarily by control of deoxidation. This control was complicated as the slab withdrawal rate increased.

Considerable porosity was encountered in all rimming-steel sections. An outer negatively segregated zone was usually characterized by a system of channel blowholes. A central porous core zone was enclosed by a ring of rounded blowholes. This core zone was definitely positively segregated. A spontaneous mechanism was suggested to explain the presence of a clearly defined core zone in the continuously cast rimming-steel sections.

This paper is another in a series intended to describe results obtained in a study of the continuous casting of rimming, semi-killed, and killed low-carbon steels. The primary objective of this programme was to manufacture usable slab sections which could be rolled into standard products and evaluated in consumer plants. It was not the direct objective of the programme to develop any fundamental concepts of the solidification process. Accordingly, the information extracted regarding solidification characteristics is incomplete and, in many aspects, admittedly speculative. Despite these limitations, this work has made it possible to contribute additional insight into those features of the continuous casting which would be important in future efforts to adapt the process to large-scale carbon-steel operations.

### Acknowledgments

The authors acknowledge the far-sighted technical policy of the Inland Steel Company for generously permitting publication of data and observations accumulated during a large-scale and costly study of the continuous casting process.

The valuable contributions of the following companies to the conduct of this work are also acknowledged: Freyn Department of Koppers Company, Atlas Steels, Ltd., Mannesmann Huttenwerke A.G.

It is obvious that a programme of such magnitude can only be completed through the combined efforts of many persons. We wish to thank all these people for their co-operation, assistance, and guidance.

### References

1. H. E. SKELLEY and R. EASTON: *Iron Steel Eng.*, 1958, May, pp. 131–142.
2. F. G. JAICKS, L. E. KRAAY, and M. TENENBAUM: *Trans AIME*, 1957, vol. 209, pp. 1057–1072.
3. M. TENENBAUM, R. N. EDMONDSON, and L. MAIR: *J. Met.*, 1958, August, pp. 431–445.
4. A. L. FIELD: *Blast Furn. Steel Plant*, 1927, vol. 45, pp. 129–136.
5. L. F. BARNHARDT, G. E. STOCK, and W. N. PORTER: *J. Met.*, 1957, August, pp. 1050–1051.
6. G. C. OLSEN: *JISI*, 1958, vol. 190, pp. 40–50.
7. K. G. SPEITH and A. BUNGEROTH: *Ibid.*, 1958, vol. 190, pp. 158–161.
8. B. M. LARSEN: *Trans AIME*, 1945, vol. 162, pp. 414–435.
9. A. HULTGREN and G. PHRAGMEN: *Ibid.*, 1939, vol. 135, pp. 133–244.
10. H. KRAINER and B. TARMANN: *JISI*, 1958, vol. 190, pp. 105–111.

ANDREW CARNEGIE SCHOLARSHIP REPORT

# The effect of rolling unstable austenitic 0·76% carbon steel at 220–300°C

*By P. R. V. Evans, Ph.D., B.Sc., and Hugh O'Neill, D.Sc., M.Met.*

SYNOPSIS

Red-hot 0·76%C steel strip was quenched in a salt bath (220–300° C) and immediately cold-rolled at reductions up to 50%. The strip was then either air-cooled ('austenrolling'), or held isothermally and cooled ('ausroltempering'). The effect of these treatments and variation of rolling speed on final microstructure and hardness was studied. The impact resistance of un-notched specimens of the austenrolled strip was superior to that of quench-tempered specimens and bainite–martensite specimens in the 625–750 VPN hardness range. A theory is proposed for this accelerated breakdown of austenite. 1532

## INTRODUCTION

NUMEROUS PAPERS have been published on the effects of deformation on the transformation of austenite. In experiments with a 0·93% carbon steel, Bish and O'Neill[1] impacted small sections quenched to metastable austenite in salt baths at 280° and 245° C. The specimens after holding in the salt bath were finally water-quenched, and it was found that a certain combination of bath temperature, holding time, and applied deformation not only gave a product of maximum hardness (e.g. 886 VPN) but this was associated with a superior value of the Meyer work-hardening index $n$. It was considered that the latter was a desirable feature so the present investigation has further explored the effects of plastic deformation on isothermal transformation. To permit the application of the results to continuous industrial processes the deformation was performed by rolling, and experiments on wire drawing in an oil bath are also being considered.

For convenience of description the two main treatments used have been defined by the following terms and are represented in Fig. 1 based on actual data:

*Austenrolling process*—Steel strip is quenched from above its critical temperature in a primary quenching bath maintained a little above its martensitic ($M_s$) temperature. The strip is held at the quenching temperature for a time insufficient to permit any isothermal transformation of the austenite and then rolled at the same temperature at the desired rolling speed. After rolling the steel is allowed to cool in still air.

*Ausroltempering process*—This follows the austenrolling process but on leaving the rolls the strip passes directly to a final holding bath which is maintained at the same temperature as the quenching bath for a controlled time. It is then either air-cooled or water-quenched as desired.

After the work had started in 1954 the possibility of obtaining improved products by these means was briefly indicated by Lips and Van Zuilen.[2] They reported improved ductility values for a Ni–Cr steel when step-quenched, rolled, and transformed to a final hardness of Rockwell C58. It then transpired that a process of step-quenching and shot-peening had also been described by Harvey.[3]

The properties of the treated steel in the present work were studied by hardness and impact tests, together with metallographic and X-ray examinations.

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Dr Evans, formerly of University College, Swansea, is now with the research department of Metropolitan-Vickers Electrical Co. Ltd. Dr. O'Neill is professor of metallurgy at University College, Swansea.

## MATERIALS AND APPARATUS

The material used was from a single coil of hot-rolled ⅛-in. basic hematite steel strip, which had been slit from 6·75 in. into ½ in. widths and had the following percentage composition:

| C | Mn | S | P | Si | Cr | Ni | N | Al |
|---|---|---|---|---|---|---|---|---|
| 0·76 | 0·52 | 0·039 | 0·026 | 0·032 | 0·25 | tr | 0·0047 | 0·053 |

The steel was kindly supplied by Messrs. J. B. and S. Lees, Ltd., and was of normal pearlitic structure. The hardness 'as received' was 330 VPN, and McQuaid–Ehn tests showed an inherent austenitic grain size of 7–6 ASTM (Fine).

The hardness by direct water- and oil-quenching was 870 and 857 VPN respectively, and the $M_s$ temperature was 200° C. The incubation period at $M_s$ was about 900 s.

The apparatus (*see* Fig. 2) comprised an inclined austenitizing furnace, horizontal primary salt-quenching bath with stirrer, 6-in. rolling mill, final isothermal holding bath (when required), and provision for final water-quenching (when required). Experiments were made to preheat the hollow rolls, but eventually unheated rolls were used, and in continuous processes they would presumably soon reach a steady temperature. In the present work 6-in. lengths of strip were treated and pulled through the train by a fine wire. A controlled drip of medium lubricating oil was allowed to enter the upper end of the austenitizing furnace and effectively prevented decarburization of the strip. Extensive control tests were carried out to see that the apparatus gave the indicated temperature

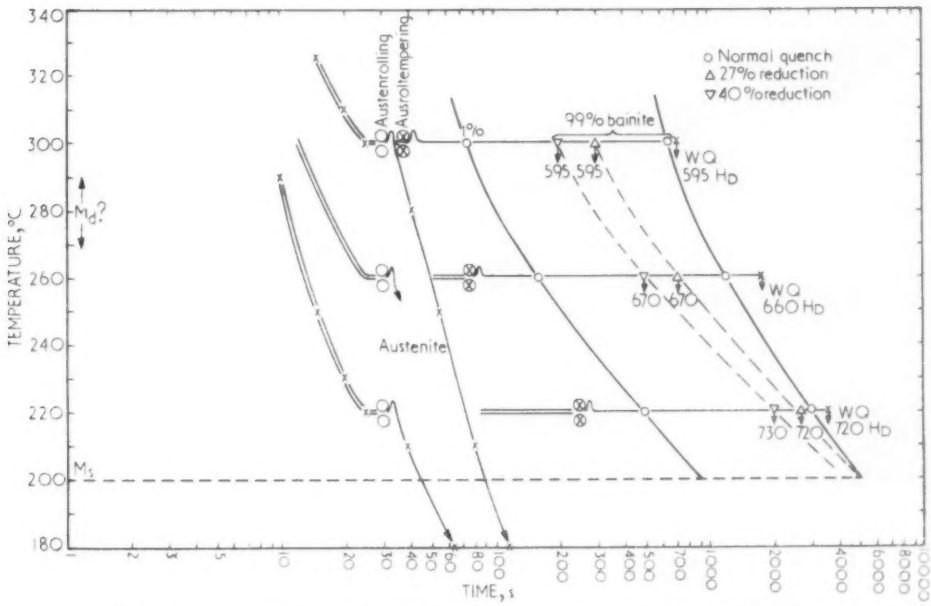



**Fig. 1—Temperature/transformation-time curves and treatment diagram**

results, and the treatment values used were as follows:

| | |
|---|---|
| Austenitizing | 830° C for 20 min |
| Primary quenching bath temp. | All three at either 220°, 260°, or 300° C |
| Rolling temp. | |
| Final holding bath temp. | |
| Holding time in primary bath (austenrolling) | 30 s |
| Holding time in final bath (ausroltempering) | 25–3600 s |
| Rolling speed 100 ft/min (except where otherwise stated) | |

Strain-gauge mill load meters were available on the rolling mill for determination of mill loads, and the reported reductions were all produced in a single pass.

Specimens for metallographic and hardness examination were cut from the centre of the strip in the direction of rolling with a 6-in. carborundum disc (1 000 rev/min) under water at a rate of 0·5 in. in 5 min. The longitudinal sections normal to the rolling surface were used for micrographs and Vickers hardness tests (20 kg load). For etching 2% nital

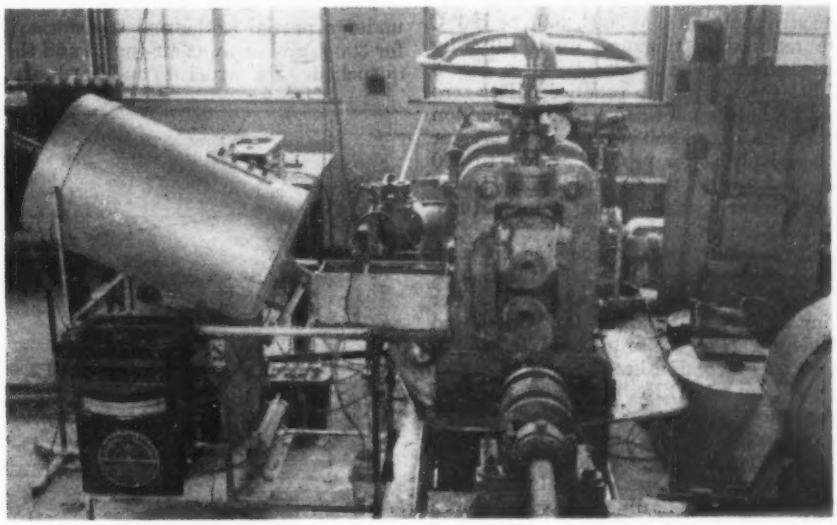
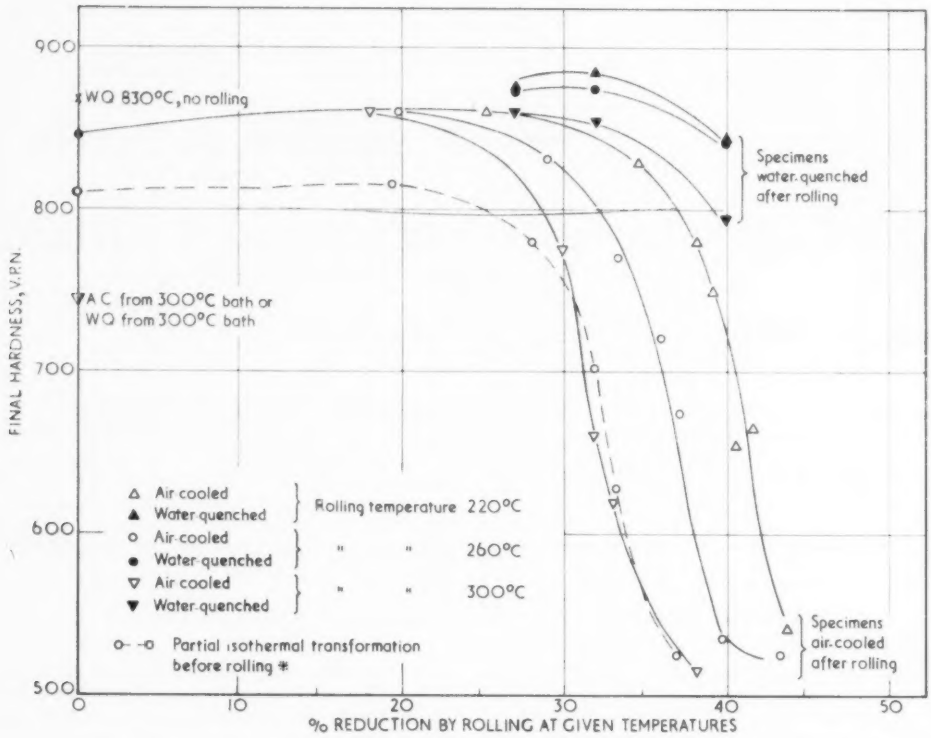


**Fig. 2—View of apparatus**




*See p. 40, 'Rolling of austenite–bainite mixtures'

**Fig. 3—Hardness values after various austenrolling treatments**

reagent was used, and the bottom edge of the micrographs was parallel to the direction of rolling. Electron-micrographs were from shadow-cast replicas.

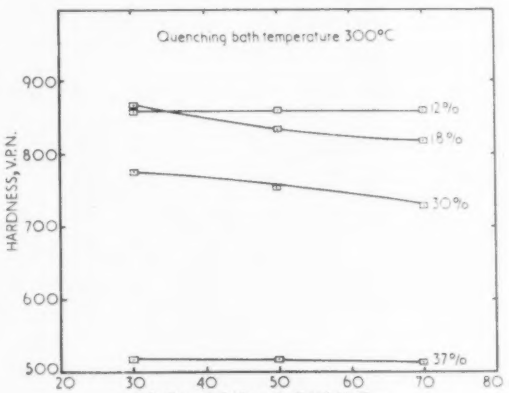
## RESULTS

### Hardness results for the austenrolling process

Austenitized strips quenched at 300°, 260°, or 220° C for 30 s were reduced in thickness up to 40% by rolling, and air-cooled. The results given in Fig. 3 show that whilst the martensite produced after zero deformation had a hardness of 845 VPN, a reduction of 20% yielded a product of 860 hardness. At greater reductions successively softer products down to 500 VPN were formed owing to strain-induced transformation, and higher quench-rolling temperatures produced a softer product.

Tests were also made by water-quenching the strip instead of air-cooling after 27%, 32%, and 40% reduction at the three rolling temperatures. For example, rolling to 32% reduction at 220° C followed by water-quenching gave a hardness of 880 VPN instead of 845. This increased hardness effect agrees with the work of Bish and O'Neill.[1] Presumably the rolling first produces strain hardening and subsequently accelerates transformation to final softer products on air-cooling, but this softening process can be arrested by immediate water-quenching and an even slightly greater general hardness may sometimes result. In the 32% austenrolled and water-quenched condition however, cracks invariably formed at the corners of the diamond pyramid indentations, and the specimens often cracked on mounting in steel clamps for preparation. The cracks generally propagated over a period of hours, originating from small surface defects in the strip.

Two pieces of unrolled steel were also allowed to undergo partial isothermal transformation at 300° C for 300 s, and one was air-cooled and the other water-quenched. The structure and hardness (745 VPN) of




Percentage reduction (R) as indicated

**Fig. 4—Effect of primary holding time**

TABLE I
Austenrolling at 260° C

| Reduction, % | Rolling speed | | | |
|---|---|---|---|---|
| | 33 ft/min | | 100 ft/min | |
| | Compression rate, in./s | VPN of product | Compression rate, in./s | VPN of product |
| 0 | 0 | 845 | 0 | 845 |
| 20·0 | 5·4 | 860 | 16·3 | 860 |
| 32·8 | 7·4 | 849 | 22·5 | 790 |
| 38·0 | 8·2 | 785 | 24·7 | 605 |

both pieces were the same, thus indicating a pronounced difference between the accelerated transformation of austenite and the austenite–bainite reaction.

The quenching-bath holding time was also shown to influence the subsequent strain-induced transformation of austenite (Fig. 4). Increasing the holding time at 300° C resulted in an increase in the extent of accelerated transformation for a given percentage reduction, (e.g. a fall of hardness at 18% and 30% reduction in Fig. 4). However, this effect was only observed when the reduction brought about partial transformation of the austenite after the minimum quenching-bath holding time of 30 s. It appears, therefore, that some internal rearrangement is in progress during the holding time that influences nucleation of the new phase. These observations are in contrast to the martensitic mode of austenite transformation where holding at temperatures just above $M_s$ tends to stabilize subsequent transformation to martensite.

### Effect of rate of strain during deformation stage

Work by Mathieu[4] and by McReynolds[5] on austenitic alloys showed that above $M_s$ increase of strain rate reduced the amount of transformation of austenite to martensite. As rate of strain may affect the austenroll process, tests were made at two rolling speeds and the results given in Table I include approximate compression rates based on Ekelund's formula.[6] It will be seen that for reductions exceeding 20% the higher rolling speed yields softer products and presumably increases the transformation rate. These findings affect application of the results of Bish and O'Neill in so far as they used high strain rates by impact.

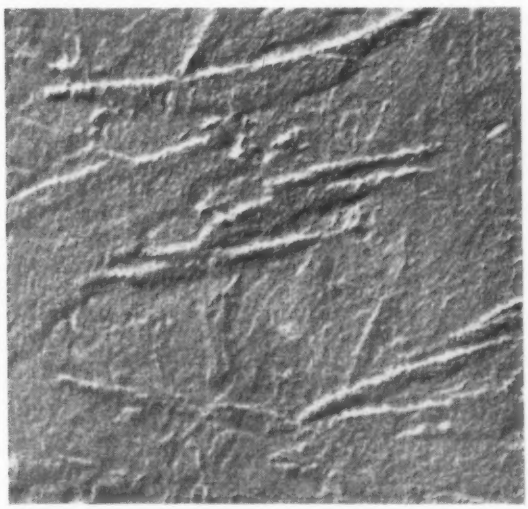


Fig. 5—Austenrolled 260° C, 0% reduction × 10 500

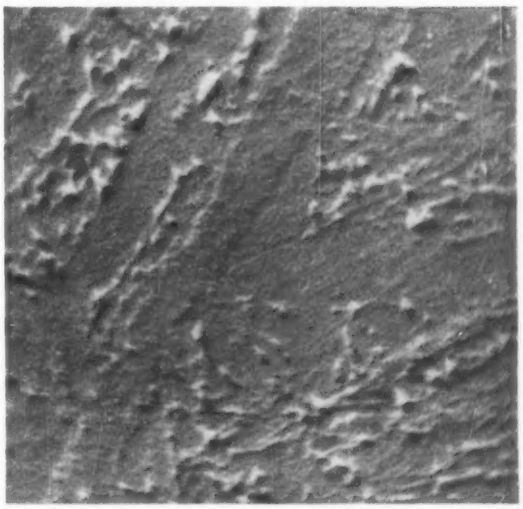


Fig. 6—Austenrolled 260° C, 19% reduction × 10 500

### Microstructure of austenrolled strip

Strip quenched at 260° C, held for 30 s, and then air-cooled had a simple martensite structure (Fig. 5). When austenrolled to 19·0% reduction the microstructure showed that a large number of the martensite needles had assumed a preferred orientation of about 30° to the direction of rolling. The corresponding electron-micrograph (Fig. 6) revealed some scattered angular nodules dispersed between the martensite plates, which themselves showed less relief effect than before. At greater reductions dark-etching grains progressively appeared, resulting in a mixture of elongated light and dark crystals. It can be seen from Figs. 7 and 8 that the dark crystals consisted of streaks of transformed constituent presumably lying along parallel (111) slip planes more favourably orientated to slip and nucleation of a new phase than those in the light grains. The latter showed faint martensitic plates at high magnifications and any twin boundaries in the original austenite seemed to facilitate the beginning of transformation.

At 40% reduction the transformation was general (Fig. 9), the structure was very fine and no austenite escaped accelerated breakdown. However, the precipitate still lay mostly along parallel lines in any one grain and was becoming discontinuous (Fig. 10). At greater reductions the precipitation product tended to spheroidize. In this range of deformation there was an appreciable fall in the rate of softening (*see* Fig. 3).

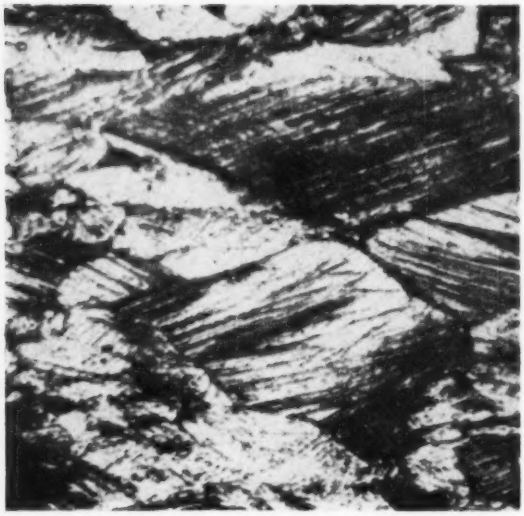


Fig. 7—Austenrolled 260° C, 35% reduction, 753 VPN × 800

Austenrolling at all temperatures usually produced a surface skin effect. At light reductions the skin was essentially a martensitic layer, but at heavier reductions it resembled material which had received much greater amounts of strain than the body of the strip.

The specimen cooling-rate between rolling and room temperature was found to have a pronounced effect on the microstructure of the final product. At reductions of 10–20%, water-quenched structures generally contained better defined and more extensive slip-plane transformation products than air-cooled structures. This suggested that in the early stages of the transformation process, quenching stresses tended to favour the accelerated breakdown of austenite.

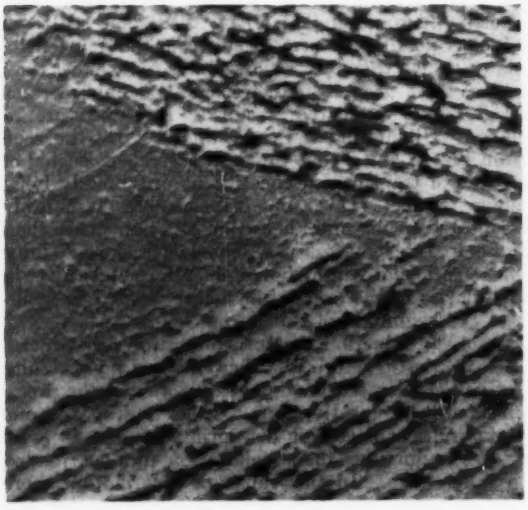


Fig. 8—As Fig. 7 × 10 500

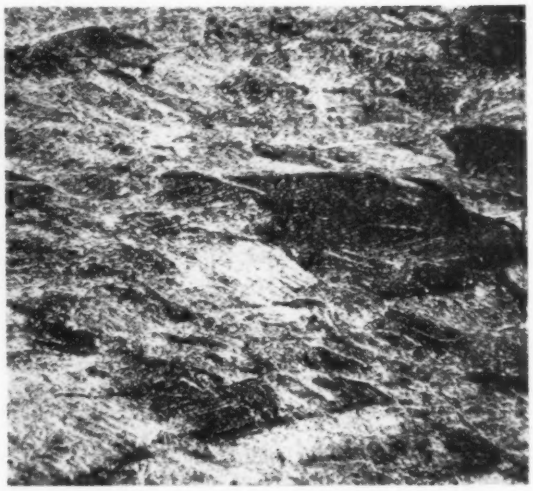


Fig. 9—Austenrolled 260° C, 40% reduction, 537 VPN × 800

**Mill load measurement during austenrolling**

It would be expected that if austenrolling at temperatures just above $M_s$ first induced rapid transformation of austenite to martensite, then the rolling-mill load would show a steep increase (e.g. ref. 7). Tests were made at 220° C with a rolling speed of 33 ft/min and reductions up to 35%. Normalized strip was also rolled at room temperature in medium lubricating oil, and after salt-bath treatment at 220° C. In the latter case it cracked at the edges. The results obtained are given in Fig. 11 and suggest that no transformation to very hard products was induced.

**Impact resistance of austenrolled products**

Impact tests were carried out with a miniature pendulum machine available in the laboratory and

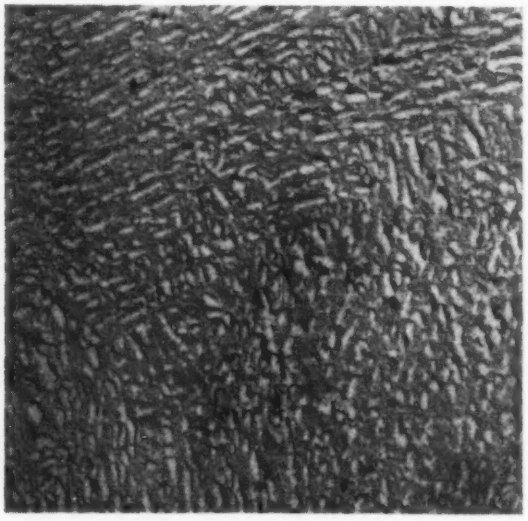


Fig. 10—As Fig. 9 × 10 500

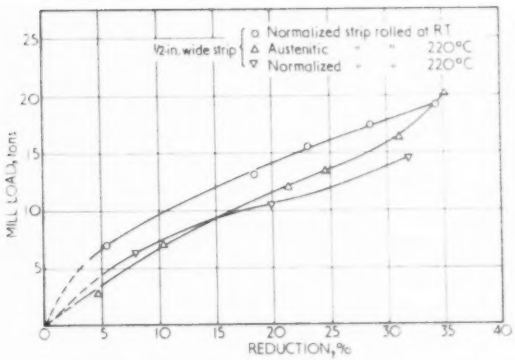



**Fig. 11—Mill loads for various strips**

giving a maximum energy of 3·6 ft-lb. After various experiments un-notched test pieces, obtained by slitting longitudinally near the centre of the strip were used. The strip was not wide enough to make impact tests in the transverse direction. The final dimensions were 1·125 in. × 0·079 in. × 0·079 in. (± 0·001 in.), the pieces being carefully ground from 0·030 in. oversize and finished by hand polishing and rubbing with 00 emery paper on the square edges. Tests were made on strip which had been given the following treatments:

(i) *Quenched and tempered* by austenitizing *in vacuo* at 830° C, followed by immediate oil-quenching and subsequent tempering in an air furnace in the range 200–400° C, followed by air-cooling

(ii) *Austempered* to 99% bainite by complete isothermal transformation at 220°, 240°, and 260° C

(iii) *Partial transformation and quenching* by bath-quenching to 300° C, holding for various times and water-quenching to give mixtures of bainite and martensite

(iv) *Austenrolling* at 300° C to various reductions to give various hardness values.

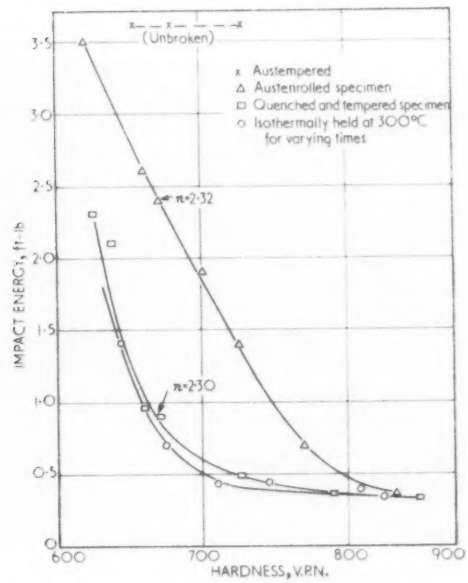



**Fig. 12—Impact resistance for various treatments**

**TABLE II**

**Diamond ball tests**

| Treatment | VPN (20 kg) | Brinell no., 30/1/15 | Meyer index, *n* | Ultimate Pu, kg/mm² |
|---|---|---|---|---|
| Austenrolled | 670 | 621 | 2·32 | 1030 |
| Quenched and tempered | 670 | 643 | 2·30 | 1005 |

The impact results are given in Fig. 12 and the austempered specimens remained unbroken in the test, and thus gave the best values. At the 800 VPN hardness level the other three processes gave about equal very low impact-resistances, but between 600 and 800 hardness the austenrolled material was superior. At 700 VPN for instance, its un-notched impact value was about four times that of either the quenched and tempered or the partially transformed specimens.

Here is evidence that austenrolled products have superior mechanical properties as compared with quenched and tempered products. If it is objected that austempered products are better still, it should be mentioned that the austempering process takes longer time as compared with austenrolling for the same high hardness values.

Tests were made with a 1-mm diamond hemispherical indenter to compare the Meyer *n* values in this hardness region. The results given in Table II show no appreciable differences for two types of material.

### X-ray examinations of austenrolled specimens

Three ½ in. × ½ in. specimens were given metallographic preparation after austenrolling at 260° C for 0%, 36%, and 41·6% reductions. X-ray diffraction photographs were obtained by grazing incidence technique, using a 19-cm Debye–Scherrer camera and $CoK_\alpha$ radiation with an iron filter. Measurement of the lines indicated martensite and retained austenite at 0% reduction, the same and possibly ferrite at 36%, whilst ferrite plus iron carbide(s) and possibly martensite were found at 41·6% reduction.

The results show that increasing deformation accelerates austenite breakdown until none was detected at 41·6% reduction. Some is retained at 36% reduction and may be a consequence of thermal and/or *mechanical stabilization* which is known to occur in austenitic steels. The following micro-indentation results suggest such a possibility.

### Micro-indentation tests of austenrolled strip

The microstructure of specimens austenrolled to partial transformation consisted in general of a mixture of elongated light-etching and dark-etching grains. The relative hardness of these was determined with a Lips-type low load tester using a Vickers diamond pyramid and a calibrated spring which applied 55·6 g. On hard specimens the readings will only be of comparative value.

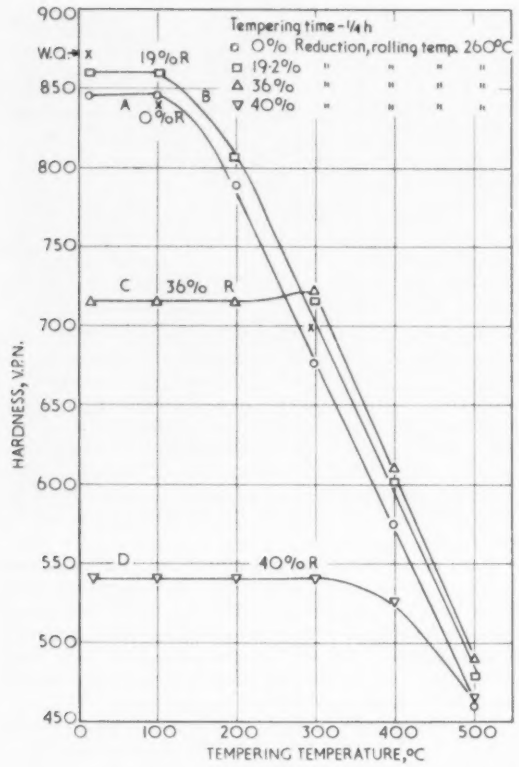



**Fig. 13—Tempering of various specimens austenrolled at 260° C**

The average of five tests on each type of grain after various treatments is given in Table III.

It is to be noted that the light etching grains (which showed no evidence of strain-induced transformation) at 220°, 260°, and 300° C nevertheless gave lower hardness values than those for the air-cooled martensite structure. Consequently it appears that these light grains have been changed by the rolling treatments.

#### Tempering of various austenrolled specimens

The stability of these air-cooled products was next studied by tempering separate specimens for (*a*) 15 min, and (*b*) 60 min at increasing temperatures, followed by air-cooling. The hardness results for various deformations at 260° C after tempering for 15 min are given in Fig. 13. The values were similar but softer after tempering for 60 min, and the general effect may be compared with the results of Holloman and Jaffe.[8]

Considering material *C* (36% reduction) before tempering, its structure consisted of light and dark grains. Any change in hardness on tempering must be due to changes in one or both types of grain. Material *D* (40% reduction) was composed of dark transformed grains before tempering and now shows virtually no change of hardness up to 300° C. It is suggested that the dark grains in *C* will behave in the same way as material *D*, and that changes in

**TABLE III**
**Austenrolled specimens**

| Rolling | | General hardness, VPN (20 kg) | Microhardness* | |
|---|---|---|---|---|
| Temp., °C | Reduction, % | | Dark grains | Light grains |
| 220 | 0 | 845 | (none) | 982 (martensite) |
| | 39·2 | 752 | 740 | 959 |
| 260 | 35·0 | 750 | 638 | 806 |
| 300 | 33·0 | 620 | 662 | 800 |

*May be about 100 units higher than Standard VPN

hardness of *C* up to 300° C are therefore restricted chiefly to the light grains.

If so, it appears that the light areas in material *C* possess different tempering characteristics up to 300° C as compared with the light martensitic material of materials *A* and *B* (0% and 19% reduction) for these show a marked softening after tempering in this range. Also, although the final rates of softening are about equal, the hardness of *A* after tempering at 300° C for 15 or 60 min is some 40 VPN points lower than *B* and *C*. These results suggest that martensite formed by cooling through $M_s$ from plastically deformed austenite exhibits retarded softening on tempering.

The observations of Wilson[9, 10] and Hundy[11] in this field may explain in an indirect way, the retarded softening noticed in the present work.

#### Rolling of austenite–bainite mixtures

Experiments were made to investigate whether the presence of bainite would influence the accelerated transformation of coexisting austenite. Specimens were held at 260° C for 360 s to produce some bainite and then rolled up to 38% reduction, followed by air-cooling. The final hardness values of the mixed

**Fig. 14—Ausroltempered 260° C, 360 s, 28% reduction, water-quenched, 833 VPN. Bainite and accelerated austenite breakdown product × 2 000**

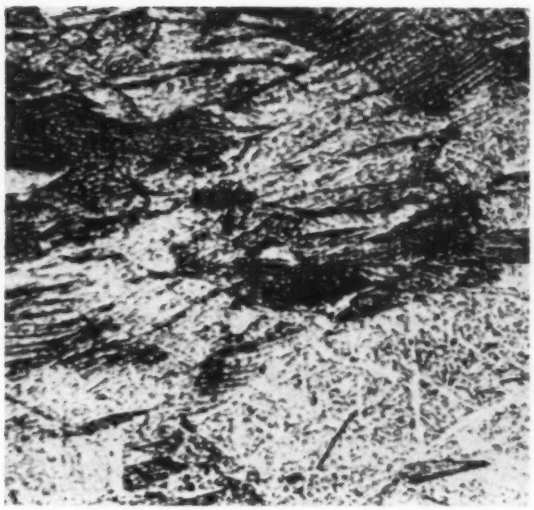


Fig. 15—Austenrolled 220° C, 41% reduction, air-cooled, tempered at 275° C for 5 s and water-quenched. Rolling speed 33 ft/min, 815 VPN × 800

Fig. 17—Ausroltempered 220° C, 27% reduction, 5 400 s, water-quenched, 720 VPN × 800

structures are included in Fig. 3 for comparison with austenrolled values. Since rapid softening occurred at about 30% reduction in the mixture compared with 35% reduction in the latter, it is considered that the austenite in the mixture is more strain sensitive.

Figure 14 illustrates bainite needles and accelerated austenite breakdown product in the same grain after 28% reduction. The differences suggest that different mechanisms are involved in their formation.

### Search for strain-induced martensite

The literature indicates that isothermal formation of martensite may be induced by plastic strain at temperatures above $M_s$ with a limiting temperature at $M_d$. To investigate the presence, if any, of strain-induced martensite, austenitized specimens were quenched at 220° C, held for 30 s, rolled to reductions

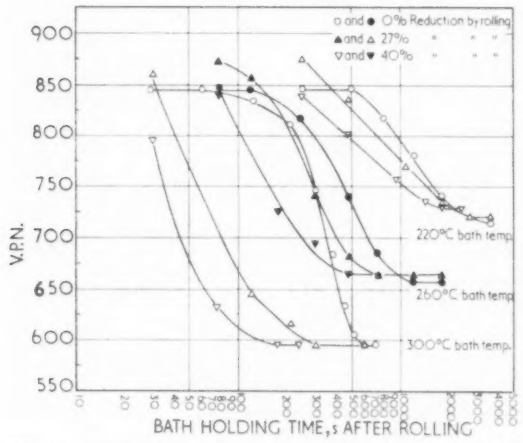



Fig. 16—Ausroltempered specimens water-quenched after various holding times

ranging from 5% to 41% at 33 ft/min, and examined by the Greninger and Troiano technique.

No darkened martensite needles were observed (e.g. Fig. 15). The light etching grains showed a specialized structure which included interlacing white plates. It should be added, however, that although the morphology of the strain-induced transformation product was dissimilar to the normal lenticular-shaped needles characteristic of martensite crystals, this reaction product may actually have formed via the martensitic mode of transformation.

### Results of ausroltempering process

The aim in these experiments was to introduce intermediate deformation by rolling during an extended isothermal holding treatment. The salt bath was therefore in two parts with the rolls in between. The rolling was applied after 37, 77, and 252 s primary holding for temperatures of 300°, 260°, and 220° C respectively, these times being considered about equal fractions of the incubation period for the three temperatures (*see* Fig. 1). The specimens were water-quenched after the final holding periods, and their hardness values are given in Fig. 16.

At each bath temperature increasing amounts of deformation progressively reduced the incubation period of transformation (as shown by the treatment times on the graphs at which softening begins), but gave the same final minimum hardness values.

These are lower (e.g. 725, 670, and 595) in proportion to the higher treatment temperatures. The observed times for the end of transformation for two amounts of reduction are plotted in the temperature/transformation time diagram of Fig. 1 and they move the finishing times to the left. These times converge towards the normal values for undeformed specimens as the treatment temperature approaches $M_s$. This differs from previously published results.[12,13]

Numerous photomicrographs have been taken of the ausroltempered products, and Fig. 17 shows almost

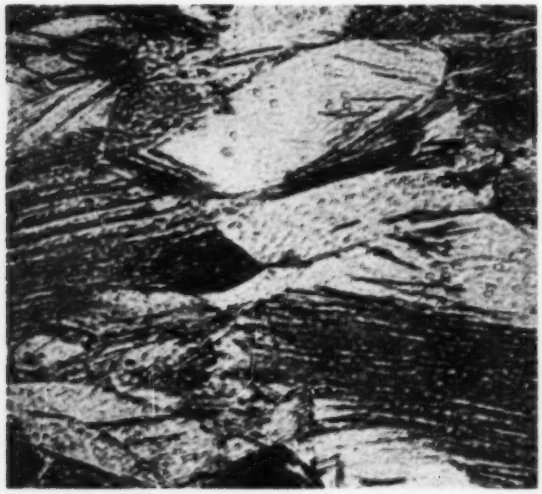


**Fig. 18—Ausroltempered 220° C, 40% reduction, 480 s, water-quenched, 800 VPN** × 800

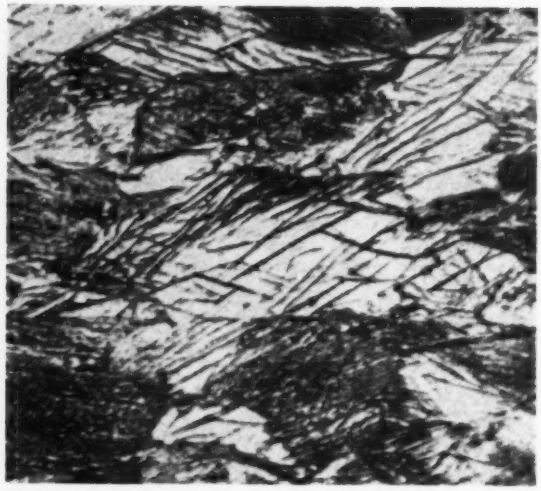


**Fig. 19—Ausroltempered 220° C, 40% reduction, 960 s, water-quenched, 763 VPN** × 800

complete transformation in the conditions given. Some series indicated a secondary transformation in austenite crystals which had not yet undergone strain-induced transformation. Thus Fig. 18 shows strain-induced product plus plain light grains. With further holding (960 s) the latter have not developed slip-plane transformation but a plate-like structure instead (Fig. 19). The plates seen in Fig. 20 after 1440 s are considered to be bainite formed by this secondary transformation which develops at holding times exceeding the incubation period of unrolled specimens. Reductions at 300° C gave less evidence of secondary transformation.

Comparison of the influence of equivalent reductions on the transformation process in austenrolled and ausroltempered strips reveal interesting effects. The results show that for all the temperatures, at 27% reduction, austenrolled strip has a higher hardness than fully transformed ausroltempered strip. Presumably this is partly because the austenrolled material contained appreciable amounts of martensite. However, at 40% reduction this hardness behaviour is reversed and most pronounced at 260° C (Table IV).

At this temperature an air-cooled austenrolled sample could be some 125 VPN points softer than a similarly treated specimen allowed to undergo complete isothermal transformation.

### Theoretical considerations of the strain-induced transformation of austenite

To account for the relatively large amount of transformation product which is rapidly formed and which nevertheless maintains a high hardness, three possibilities may be considered:

(1) Strain-induced martensite may form along shear planes in the parent austenite and immediately precipitate carbide due to the localized increase in temperature by rolling. However, the experimental results showed that an increase in bath holding time and/or speed and temperature of rolling produced more austenite breakdown. This evidence suggests that transformation was not of the normal martensitic type.

Indeed, consideration of the crystallographic features of plastic deformation in austenite and of martensite formation in high-carbon steels might suggest that the production of strain-induced martensite is not possible. This is in contrast with low-carbon austenites where the factors appear favourable and where experimental work has established the existence of a strain-induced martensitic transformation.

(2) The formation of strain-induced bainite from austenite is also thought to be improbable because

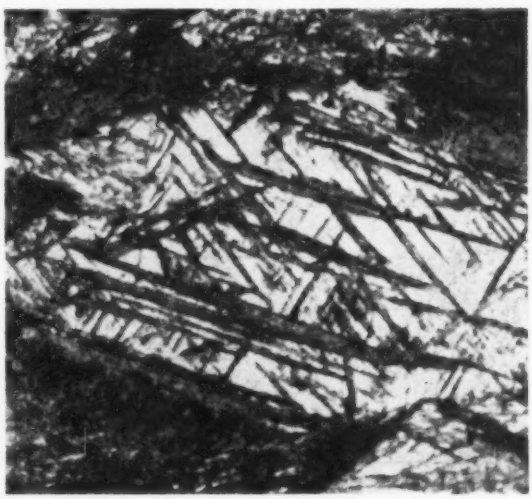


**Fig. 20—Ausroltempered 220° C, 40% reduction, 1440 s, water-quenched** × 2000

the product was sometimes harder than normal martensite and its microstructure was different from bainite. The influence of cooling rate on accelerated transformation was also in contrast to that of the bainite reaction.

(3) The present authors suggest that the accelerated breakdown of austenite in the present work occurs in two main stages as follows:

$$\text{Austenite} \xrightarrow[\text{work}]{\text{Cold}} \underset{+}{\epsilon\text{-Carbide}} \underset{\text{Martensite}}{} \xrightarrow{\text{Diffusion of Carbon}} \underset{+}{\text{Cementite}} \underset{\text{Ferrite}}{}$$

The views of Bilby and Christian[14] on the martensitic transformation of $\alpha$ cobalt (f.c.c.) $\rightarrow \beta$ cobalt (h.c.p.) and of Cina[15] in Fe–Ni–Cr alloys are a guide to the proposed process.

Jack[16] first established the formation of h.c.p. $\epsilon$-carbide ($Fe_2C$–$Fe_3C$) during the tempering of martensite. He suggested that the (101) $\epsilon$-carbide planes are parallel to and coherent with the (101) martensite planes. Also, that the (101) planes of the martensite are in turn parallel with the (111) shear planes of the parent austenite.

Plastic deformation produces dislocation ' pile-ups ' in the parent phase resulting in strain hardening. During this process stacking faults are generated along (111) austenite planes and nuclei of a second phase having an h.c.p. structure form directly from these structural imperfections. These atomic movements could be influenced by the presence of preferred sites which may exist in the parent phase before deformation. The influence of quenching-bath holding time before deformation on the accelerated transformation process indicates such a possibility. The formation of stacking faults by the dissociation and interaction of dislocations in f.c.c. structures has been outlined by Cottrell,[17] whilst Barrett[18] has shown experimentally that plastic deformation can produce such faulting. It is suggested that this second phase is $\epsilon$-iron carbide and its growth occurs with the aid of thermal activation. The formation of $\epsilon$-iron carbide would tend to relieve strain-hardening effects in the austenite matrix. However, in those regions where accelerated breakdown did not occur the strain-hardened austenite transformed to martensite on cooling to room temperature.

The $\epsilon$-iron carbide is visualized as a metastable transitional phase being the product of what is termed martensitic nucleation.[19] During the initial stages of formation of the $\epsilon$-carbide the surrounding matrix transforms to martensite on cooling with which the carbide may be coherent. The appreciable amounts of accelerated transformation product that were observed metallographically in specimens water-quenched immediately after rolling suggest that the rate of nucleation is high, the necessary atomic movements are small, and long-range diffusion is probably not of great importance, at least, in the early stages of the process.

Thus, during the onset of accelerated breakdown it is thought that the finely dispersed transitional $\epsilon$-carbide and martensite are strained into precise atomic registry or coherency, giving rise to coherency hardening which is common amongst age-hardening alloys. The high degree of localized strain caused by the formation of $\epsilon$-carbide, and manifested as an increase in broadening of X-ray reflections, has been recorded by several investigators.[20] These considerations would account for the high initial hardness values recorded during ausroltempering. It is interesting to note in this connection that Roberts, Averbach, and Cohen[21] established that precipitation of $\epsilon$-carbide resulted in a hardening effect during a study of the kinetics of the first stage of tempering.

TABLE IV

**Rolling (and holding) temperature 260°C**

| Treatment | Reduction, % | Final hardness, VPN |
|---|---|---|
| Austenrolled | 27·0 | 845 |
| Ausroltempered (720 s) | 27·0 | 660 |
| Austenrolled | 40·0 | 535 |
| Ausroltempered (500 s) | 40·0 | 660 |

The $\epsilon$-iron carbide then grows by a diffusion controlled process resulting in the particle size increasing, and on exceeding a critical value, the precipitate breaks away from the matrix with an accompanying fall in hardness. During this stage, the metastable $\epsilon$-iron carbide transforms to the more stable cementite, i.e. nuclei of the cementite structure are obtained by only small distortions of the transitional structure. The growth of $\epsilon$-iron carbide may occur during air-cooling or on isothermal holding after rolling.

The final stage of the transformation to softer products is a result of spheroidization and coalescence of the cementite particles. A similar state of affairs exists in the third stage of tempering of quenched iron–carbon alloys.[22]

Previous results infer that the induced transformation is merely an acceleration of the austenite $\rightarrow$ bainite reaction. The present findings may perhaps be explained by the relatively high strain-rates employed, giving high localized temperature increases. Furthermore, Christian[23] pointed out in a qualitative way how strain-rate might influence the reflection of half-dislocations in f.c.c. crystals and suggested that high strain-rates favoured the generation of twins. It is thought that high strain-rates might also help in the generation of stacking faults in the parent phase and consequently aid the accelerated transformation process.

The assumption that carbide precipitation marked the onset of austenite breakdown may be viewed in the light of the quasi-equilibrium diagram of the bainite reaction proposed recently by Krisement and Wever.[24] The localized heating at slip planes, occurring during deformation by rolling, might increase the temperature of the 0·76% carbon austenite in the vicinity of these planes above his Z-curve, i.e. into region III. In this region Krisement and Wever suggested that free energy considerations favoured the breakdown of austenite by initial carbide precipitation. This view lends support to the present theory that the accelerated transformation of austenite could begin with the precipitation of a carbide.

The favourable impact resistance of the hard austenrolled products (Fig. 12) may be due to their low internal micro-stresses as a result of the proposed

transformation process. The microstructure of elongated light and dark grains may also contribute to the improved results.

## CONCLUSIONS

(1) The rolling of unstable austenitic 0·76%C steel at 220–300° C produced a slightly harder final product for reductions up to 20%. In general it accelerated the rate of austenite breakdown. Above 27% reduction the incubation period vanished and increased deformation up to 40% reduction increased the amount of transformation and softening.

(2) Increase of rolling speed increased the extent of transformation.

(3) Austenite breakdown first occurred in those crystals most favourably orientated for slip by rolling, whilst it was delayed in others. The product therefore consisted of elongated 'hard' and 'softer' grains.

(4) The strain-induced transformation product did not resemble normal bainite. Strain-induced martensite was not detected.

(5) Tempering of austenrolled specimens suggested that the martensite formed from strained austenite exhibited retarded softening.

(6) The longitudinal (un-notched) impact properties of austenrolled material in the 625–750 Vickers hardness range were superior to those of quenched–tempered and martensite–bainite material.

(7) It is suggested that the accelerated breakdown of austenite by strain to give streaked dark grains takes place by martensitic nucleation of $\epsilon$-iron carbide followed by its growth to a more stable cementite.

### Acknowledgment

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### References

1 G. R. Bish and H. O'Neill: *JISI*, 1953, vol. 173, pp. 398–405.
2. E. M. H. Lips and H. Van Zuilen: *Met. Prog.*, 1954, vol. 66, Aug., pp. 103–104.
3. R. F. Harvey: *Iron Age*, 1951, vol. 168, Dec. 21, pp. 70–71.
4. K. Mathieu: *Arch. Eisenhütt.*, 1942, vol. 16, pp. 215–218.
5. A. W. McReynolds, *J. Appl. Phys.*, 1946, 17, 823–833.
6. J. Ekelund: *Steel*, 1933, vol. 93, Aug. 21, pp. 27–29.
7. T. Angel: *JISI*, 1954, vol 177, pp. 165–174.
8. J. H. Hollomon and L. D. Jaffe, *Trans AIME*, 1945, vol. 162, pp. 223–247.
9. D. V. Wilson: *JISI*, 1954, 176, p. 28.
10. D. V. Wilson: *Acta Met.*, 1957, vol. 5, pp. 293–302
11. B. B. Hundy (in discussion): *JISI*, 1953, vol. 174, p. 345.
12. F. C. Thompson and M. D. Jepson: *Ibid.*, 1949, vol. 162, pp. 49–56, or 1950, vol. 164, pp. 27–35.
13. S. Bhattacharya and G. L. Kehl: *Trans ASM*, 1955, vol. 47, pp. 351–377.
14. B. A. Bilby and J. W. Christian, in symposium: "The Mechanism of Phase Transformation in Metals," pp. 121–172: 1955, Institute of Metals, Monograph No. 18.
15. B. Cina: *JISI*, 1954, vol. 177, pp. 406–422.
16. K. H. Jack. *Ibid.*, 1951, 169, pp. 26–36.
17. A. H. Cottrell: "Dislocations and Plastic Flow in Crystals": 1953, Oxford, Clarendon Press.
18. C. S. Barrett, *Trans AIME*, 1950, vol. 188, pp. 123–135.
19. J. S. Bowles and C. S. Barrett, in "Progress in Metal Physics," (Ed. Chalmers), 1953, vol. 3, pp. 1–41.
20. J. A. Wheeler and M. A. Jawson: *JISI*, 1947, vol. 157, pp. 161–166.
21. C. S. Roberts, B. L. Averbach, and M. Cohen, *Trans ASM*, 1953, vol. 45, pp. 576–599.
22. E. D. Hyam and J. Nutting: *JISI*, 1956, vol. 184, pp. 148–165.
23. J. W. Christian: *Proc. Roy. Soc.*, 1951, vol. 206A, pp. 51–64.
24. O. Krisement and F. Wever, in symposium: "The Mechanism of Phase Transformation in Metals," pp. 253–263: 1955, Institute of Metals, Monograph No. 18.

# A modified bend test for hardened tool steels

*By G. Hoyle, B.Sc., L.I.M., and E. Ineson, B.Sc., F.I.M.*

## INTRODUCTION

THE TESTING of hardened tool steels with a view to assessing and comparing their properties is a subject of considerable complexity in comparison to the testing of most other types of steel. The reasons for this are to be found in the nature of the steels themselves and in the fact that they are used in an extremely hard condition. The ultimate guide to the quality of any tool steel must be found in the results of cutting tests performed under service conditions, but such tests are fraught with difficulties if all factors are to be kept under strict control. It is often difficult to be sure that a cutting test is not in fact a measure of the efficiency of sharpening the tool or a measure of the vibration of the machine, or even the heterogeneity of the test log being cut by the tool. Cutting tests are, moreover, tedious and time-consuming and involve a

## SYNOPSIS

A bend-test apparatus has been specially designed for the testing of fully hardened tool steels at atmospheric or elevated temperatures. A method is described for calculation of stresses, which makes adequate correction to classical bending formulae to enable their application to specimens behaving in a non-elastic manner. Analysis of the results of several series of experiments is made, and hypotheses are advanced for relating these properties to the 'toughness' of tool steel, a quantity which remains undefined. Various parameters are given which, it is suggested, assist in the appraisal of tool steels, whether of the plain carbon, low-alloy, or high-speed type. 1596

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The authors are on the staff of the Division's Metallurgy section.

considerable wastage of material. It seems obvious then, that simple laboratory tests, even if only intended as preliminary sorting tests, would have some value in assessing the properties of hardened tool steels having a diamond hardness of the order of 800 or more.

In service, tools may fail either by wear or by fracture. It is the latter method which is under consideration in this paper and for the want of a better term the ability of a tool to resist failure by this method is referred to as ' toughness,' a property which appears to be rather vague and which has so far eluded all attempts to define it in precise terms.[23] Whilst this latter statement still holds, it is hoped that the present paper will make a contribution to a better understanding of the term ' toughness.'

Numerous workers have undertaken to examine the problem and have used a variety of tests such as static torsion,[1-4] impact torsion,[4-10] un-notched impact,[11] repeated blow tests,[9-14] fatigue tests,[15] compression,[4] and bend tests.[4, 16-21] Many of these have been reviewed by Clayton-Cave and Ineson.[22] Workers in this field have met with varying degrees of success, but the results given in the literature are not sufficiently conclusive to make possible the selection of any single test as being adequate to provide the best information on the toughness of tool steel. Geller[4] has stated that the life of a tool will be enhanced if it is heat-treated to give the highest mechanical properties for a given hardness. Emmon[2] using the static torsion test, quoted values for a ' toughness coefficient ' which represented the product of ultimate torque and ultimate deflection. Knight[10] has developed a ' performance rating ' for armour-piercing shells, on the basis of torsion impact test results, which is proportional to the product of the torsion impact energy and the difference between the hardness of the specimen and the minimum acceptable hardness for the particular application. Grobe and Roberts[22, 23] have defined toughness as a combination of the two factors, (*a*) ability to deform before breaking (ductility), and (*b*) the ability to resist permanent deformation (elastic strength).

A survey of some of the tests referred to is at present being carried out by members of the BISRA Tools Steels Committee, and no attempt will be made in the present paper to compare the merits or otherwise of the various methods. In the BISRA laboratories much attention has been paid to the development of the bend test for testing hardened tool steels, and the object of the present paper is to describe this development and to present some results which indicate the potentialities of the method of testing.

## THE BEND TEST

Early work in connection with the bend testing of tool steel was carried out by Barry,[16, 17] d'Arcambal,[18] Amberg,[19] and others. A major contribution to the development of the test was made by Grobe and Roberts,[20] who examined in detail the relative merits of 3- and 4-point loading (in U.S. terminology single- and double-point loading), and also carried out a great deal of work to determine the optimum specimen size. They finally recommended the 4-point loading system, and found that specimen size did have an effect on the results. The method was subject to some considerable scatter, as are all results on high-speed and other tool steels, but by statistical methods it was shown that reasonably accurate measurements of strength and ductility could be made, the properties considered being yield strength (sometimes referred to as limit of proportionality), bend strength (breaking stress), total deformation, and plastic deformation.

Geller[4] has stated that the scatter in results of bend tests on quenched and tempered tool steels is of the order of 6%, i.e. somewhat greater than that for the torsion test, but nevertheless he found that the bend test was capable of detecting changes in ductility due to variations in composition, structure, heat treatment, and carbide heterogeneity. He considered that the bend test alone was inadequate for a complete appraisal of a tool steel, owing to " non-uniform distribution of stress over the cross-section and length of the test piece," and suggested the use of the torsion test in addition. The inference from his paper is that he was using a 3-point loading system, and some of his objections may therefore not apply to the 4-point system, in which there is a uniform bending moment over the central span.

## DESIGN OF APPARATUS

After consideration of the problem and the kind of commercial apparatus available it was decided that for the present work it was preferable to design an instrument specially for the testing of tool steels, and to incorporate certain features considered to be of special interest and value in the testing of high-speed steels. Also, the instrument was designed so as to enable a small specimen to be used: this was thought to be desirable so as to conserve material under test and also to make the removal of test pieces from a wide range of finished components possible if necessary.

In the apparatus ultimately developed the main departure from the conventional design was to enable tests to be made at temperatures other than room temperature, in case it was required to evaluate the properties of high-speed steels or die steels at their normal working temperatures. So far as is known, all previous devices for bend testing have involved the use of some jig in which the specimen under test was surrounded by vital parts of the apparatus, and the incorporation of a furnace to heat specimens under such conditions presented considerable difficulties. Provision has therefore been made in the present apparatus to accomodate the test piece at one end of the machine, with all load and deflection-measuring instruments sufficiently far removed from the specimen so that it could be immersed in a liquid heating bath during the test.

The prototype equipment is shown pictorially in Fig. 1, and comprises the test gear itself and a galvanometer, together with a d.c. power pack (not shown), the latter two items being used in connection with the load measurement. The principle of operation is shown in Fig. 2. A steel plate and four steel rods form a frame and hold a block in which a central hole is machined to accomodate an inner sliding block which in turn is connected to a lead screw by an elastic link fitted with strain gauges. The load is applied by hand operation of the worm gear. Two pairs of $\frac{1}{8}$-in.

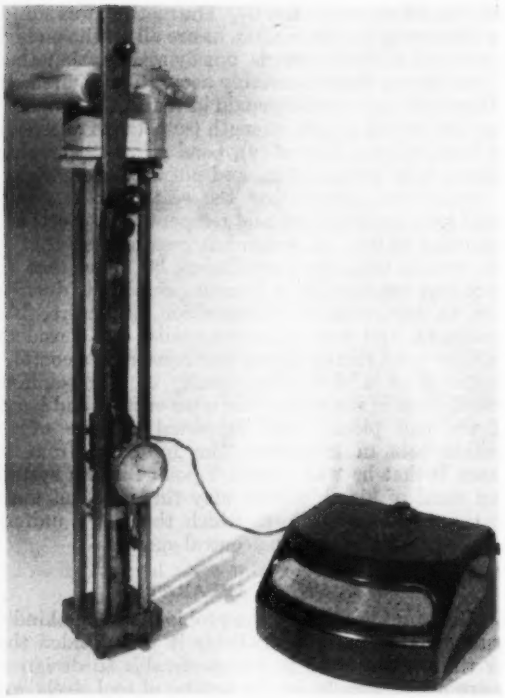


Fig. 1—Prototype equipment

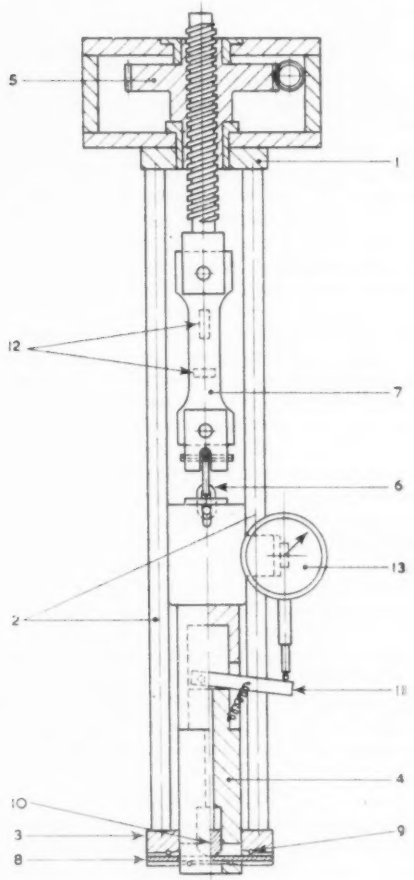
dia. tungsten carbide rods are fitted into the blocks and constitute the bearing points for the specimen giving a load distribution as shown in Fig. 13. A light plunger, which is free to move in an axial hole drilled in the inner block, rests on the specimen at its mid point; the other end bears on a lever, the fulcrum of which is a fixed point on the inner block. Thus any deformation of the specimen relative to the two central supports is registered on the dial gauge which is actuated by the lever. Calibration of the system is carried out by direct measurement using a depth micrometer.

The sensitivity of the measuring system is such that one scale division on the dial gauge is equivalent to 0·00014-in. deflection of the specimen, and interpolation can be made to 0·2 of a scale division if necessary.

The elastic link is constructed from 1%C–1%W hacksaw sheet steel, and the dimensions have been so chosen that the strip behaves in an elastic manner when subjected to a tensile load of up to $\frac{1}{2}$ ton, which, with the particular design of apparatus and test piece, gives a maximum attainable nominal stress on the specimen of over 200 tons/in$^2$. Strain gauges are affixed to the strip and are connected in the bridge network shown in Fig. 3. The power supply for this circuit is obtained from an Ediswan stablized power unit, type R 1095, which is normally adjusted to give an output of 120 V d.c. The out-of-balance current is measured by a spot galvanometer. After calibration in a tensile testing machine, the link is used in the bend-test apparatus to measure the load applied to a specimen. To prevent errors due to the slight twisting of the link, which is possible owing to the particular method of applying the load, a short length of chain has been incorporated in the transmission system to accomodate the torsion.

**Proposed alternative design**

Having established the usefulness of the prototype apparatus described above, an improved model is proposed incorporating motorized drive and rather more robust extension and load measuring devices: this is at present being designed. In addition, an alternative unit is being designed which may be used with the Hounsfield tensometer, probably for room temperature tests only. It is expected that this unit will be so constructed that a stress/strain curve may be plotted on the standard tensometer recorder.




1. Steel plate
2. Steel rods forming frame
3. Block
4. Inner sliding block
5. Worm gear
6. Chain
7. Elastic link
8. Specimen
9. Bearing points
10. Light plunger
11. Lever
12. Strain gauges
13. Dial gauge

Fig. 2—Diagram of prototype equipment

## SPECIMEN SIZE AND PREPARATION

As a result of much exploration the optimum specimen size was found to be 0·1 in. × 0·2 in. × 3 in. The machine is capable of accomodating pieces of slightly differing dimensions, but if the thickness is increased the amount of stress which can be applied is insufficient to fracture some specimens, whereas the amount of deformation undergone by thinner samples is often too great for measurement by the particular system adopted.

Specimens, after heat treatment, are prepared by surface grinding, using a copious supply of soluble oil. Extreme care is necessary at this stage, especially with samples in any but the fully tempered condition, and even then it is easy to ruin a specimen. All grinding is done so as to leave the grinding marks parallel to the axis of the specimen. Grobe and Roberts[20] used dry grinding in the preparation of specimens, but here also the need for great care was emphasised.

Tarasov[24] has given an exhaustive treatment of the subject of grinding, particularly of high-speed steel, and concludes that dry grinding adversely affects the surface properties only if the depth of feed exceeds 0·001 in. Wainwright and Nickols[25] have shown by X-ray methods that freedom from surface changes is possible when grinding gauge steels (which are presumably more sensitive to 'burning' than high-speed steels) by using the dry method with a feed of not more than 0·0005 in.; on the other hand rough grinding (0·003-in. feed) affected the surface to a depth of 0·002 in. It may appear, therefore, that the use of a coolant is an unnecessary precaution in the grinding of bend-test specimens. It is proposed to undertake experiments to examine this aspect more fully in the future.

## TESTING PROCEDURE

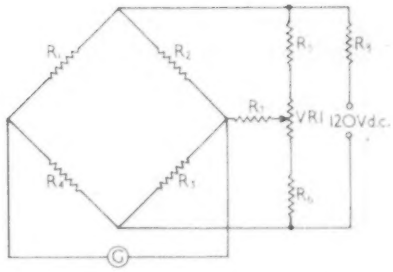
The normal method of operating the test is to take dial gauge readings (deflection) at equal increments of load (galvanometer readings) until fracture takes place. It has been found advisable to wrap a cloth or cotton wool round the specimen during test, as fracture often takes place with considerable violence and fragments of specimens may be projected with high velocity.

Elevated-temperature tests are conducted by immersing the lower end of the test apparatus in a heated oil bath, or other suitable liquid bath, taking care to avoid heating of the remainder of the apparatus. Satisfactory tests have so far been made up to 300° C, but no results are given here. There seems to be no objection to extending the use of the apparatus to testing at sub-zero temperatures, if the need arises, but no trials have yet been made.

## CALCULATION OF RESULTS

From the graph of galvanometer reading $G$ against dial gauge reading the following values are obtained:

(i) $G_L$: galvanometer reading at limit of proportional bending, i.e. the point at which the graph deviates from linearity
(ii) $G_F$: galvanometer reading at fracture
(iii) $\lambda_L$: dial reading corresponding to $G_L$
(iv) $\lambda_F$: dial reading corresponding to $G_F$.




$R_1$, $R_2$: Strain gauges for tension (longitudinal)
$R_3$, $R_4$: Strain gauges for compression (transverse)
$R_2$, $R_4$: 20K $R_5$: 66·7K
$R_6$: 30K VR1:-20K Potentiometer

**Fig. 3—Bridge network**

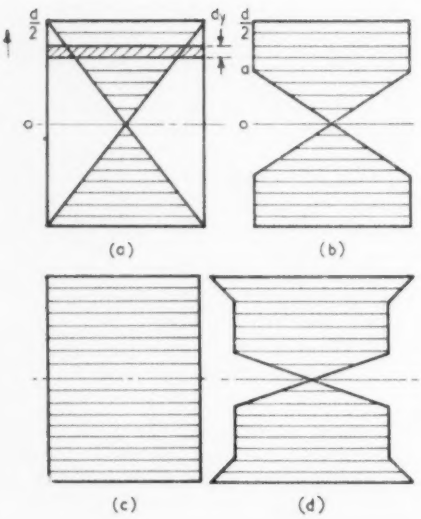
This information, together with the dimensions of the specimens and the experimentally determined calibration constants for stress and deflection measurements, is sufficient to enable the evaluation of the major properties of the specimen by classical bending theory. (Details of calculation are given in Appendix I.) The values obtained are:

(i) Fracture stress, tons/in²
(ii) Limit of proportional bending, tons/in² (sometimes called yield stress)
(iii) Total deflection, in.
(iv) Plastic deflection, in.
(v) Modulus of elasticity $E$, lb/in².

## ERRORS INVOLVED BY USE OF CLASSICAL BENDING THEORY

There are several basic assumptions which are inherent in the formulae derived from classical bending laws. The most important of these is that the material behaves in an elastic manner throughout, and this is certainly not true in the bend test once the limit of proportional bending has been exceeded. This has been recognized by all who have used the test, but it has been common practice to ignore this and to use the uncorrected laws to obtain a 'nominal' stress evaluation in the plastic region.

Hamaker, Strang, and Roberts,[21] conducted a series of experiments to determine the relationship between nominal bend breaking stress and tensile strength, and quoted ratios of bend/tensile properties for two selected grades of tool steels. According to their results, for specimens heat treated to a hardness level within the range 30–55 Rockwell C, tensile strength is equivalent to 45–50% of the bend breaking stress, and the tensile limit of proportionality is equivalent to 72% of the limit of proportional bending. The discrepency in the elastic region is somewhat surprising, and it is interesting to note the result obtained if it is assumed that this error is continued in the plastic region. The value of nominal breaking stress (bend) would then be reduced by 28%, and this corrected value than becomes approximately 1·5 times the tensile strength. Geller[4] has used, in effect, a correction factor equal to 1·5 when dealing with the non-elastic deformation of ductile materials. He points out, however, that the value of the correction factor is dependent on the material and




(*a*) Ideal case, no plastic deformation
(*b*) Partial plastic deformation
(*c*) Ideal case, full yielding
(*d*) Partial yield, partial work hardening

**Fig. 4—Schematic stress distributions in bend-test specimens**

varies between 1·0 and 1·5 according to the degree of plastic deformation at the fracture point. Roark[26] has given details of empirically determined constants for use with various materials.

The method by which the constant 1·5 is obtained is discussed theoretically in Appendix II. The basic assumption is that at the limit of proportionality the stress distribution through the specimen is ideal and linear, as shown in Fig.4*a*. Once the outer fibres have yielded, it is assumed that the surface stress remains constant whilst fibres progressively nearer to the neutral axis reach a similar stress (Fig. 4*b*) and finally all fibres yield (Fig. 4*c*).

Appendix III gives a method for calculating the correction factor if a stress distribution as in Fig. 4*b* is assumed, and by its use individual points on the bend-test curve can be corrected. Figure 5*a* shows typical uncorrected and corrected bend-test curves, and Fig. 5*b* gives, for comparison, a tensile curve for high-speed steel which is typical of those obtained by Hamaker *et al.*[21]

It is apparent from Fig. 5*a* that the corrected stress does not remain constant after yield has begun. This suggests that work-hardening is taking place at the surface, and that a more correct picture of the true stress distribution would be given by Fig. 4*d*. Work-hardening, to the extent of about 20 points on the Vickers scale, has been measured on the tension face of bend-test specimens of high-speed steel after test. The method in Appendix III is, therefore, still subject to some error, but mathematical treatment of stress distributions of the type shown in Fig. 4*d* is extremely cumbersome. A further basic assumption in determining the corrected stress/strain curve is that the neutral axis remains at the centre of the test piece, and that the stress distribution is symmetrical about the neutral axis. In spite of these errors, it is considered that the proposed method gives results which are substantially more accurate than those given by unmodified theory. A limited number of tensile tests has been possible on fully hardened and tempered 18–4–1 steel; five specimens gave a mean tensile strength of 145 tons/in$^2$ and the mean corrected breaking stress computed from ten bend specimens, similarly heat treated, was 149 tons/in$^2$. This seems to support the validity of the arguments advanced.

## INTERPRETATION OF RESULTS

As mentioned earlier, from the uncorrected bend-test curve, the following properties can be found:

(i) Limit of proportionality, $L$
(ii) Nominal breaking stress, $B$
(iii) Total deflection
(iv) Plastic deflection
(v) Modulus of elasticity, $E$.

From the corrected bend-test curve the following additional properties can be determined:

(vi) ' True ' breaking stress, $\phi$
(vii) Energy to fracture, $U$.

In determining $\phi$ it is necessary to calculate the extent to which yielding has taken place and a further quantity is therefore known:

(viii) Percentage of total specimen yielded at fracture, (*see* Appendix III).

An obvious additional factor of importance in interpreting the results in relation to properties is:

(ix) Hardness, measured by the Vickers test, $H$.

It is now necessary to examine these properties and to analyse their individual and cumulative effects on the property of ' toughness ' or ' quality.' The validity of some of the arguments advanced is demonstrated in a later section by experimental results.

### Hardness and limit of proportionality

It seems reasonable to assume that a high value of limit of proportionality is a desirable feature in any tool, as stated by Grobe and Roberts, and likewise it appears obvious that a tool must have adequate hardness for its particular application. These two criteria appear to be related.

Hardness may be defined as resistance to plastic deformation, and Williams[27] has pointed out that two of the main factors involved in the equation for resistance to penetration (i.e. indentation hardness) are the same as those which come into operation when

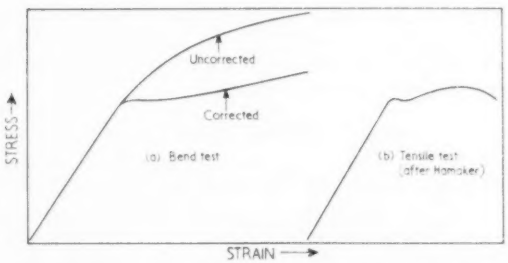



**Fig. 5—Comparison of form of bend and tensile curves for high-speed steel, not to scale**

a body under tension reaches the yield point, namely (*a*) the mechanical tearing apart of the crystals, and (*b*) the frictional resistance of the particles to slip (internal friction). The inference drawn by Williams is that, if practicable, the yield point or tensile strength of a substance would be a better measure of hardness than any other property. Normally the indentation test is the most convenient method of measuring hardness, and it has been common practice to use such measurements as a guide to tensile properties (at least for low-alloy steels).

Hamaker *et al*[21] found that for adequately tempered high-speed steels there was a smooth relationship between Rockwell hardness and limit of proportionality in bending. For inadequately tempered specimens, however, the limit of proportionality was below that obtained from their graph. The present authors have found a similar relationship using Vickers hardness (*see* Fig. 6).

The points on this graph are derived from the results of tests on a wide range of steels, ranging from plain carbon to the most highly alloyed high-speed steels, and including samples treated in many different ways. Some of the heat treatments used were deliberately designed to give samples in a condition unsatisfactory for service use, although most of the cycles used are known to have been used commercially at some time. The points in Fig. 6 corresponding to specimens which it is certain have been properly heat treated to give the best combination of properties, in regard both to service behaviour and bend-test properties, are identified and the mean values computed from six such batches of specimen have been used to draw the straight line. It appears, therefore, to be established that for a given hardness $H$ there is an optimum value $\theta$ for limit of proportionality, which is not attained if the heat treatment is inadequate. It is suggested therefore that the ratio $L/\theta$ might be used as a parameter expressing the fraction

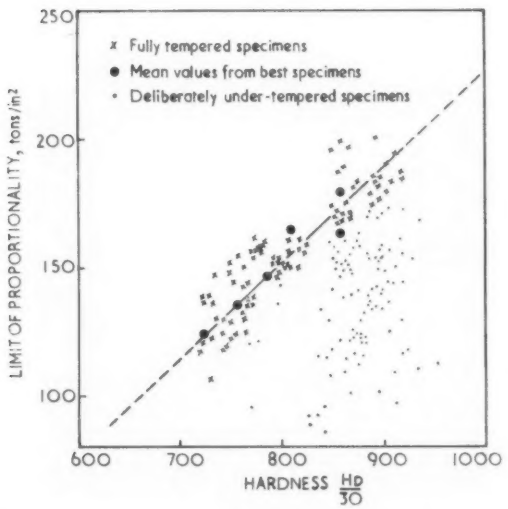



**Fig. 6—Relation between hardness and limit of proportionality**

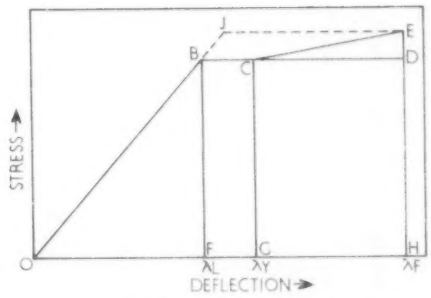



**Fig. 7—Hypothetical corrected bend-test diagram. Effect of various factors on energy to fracture**

of the maximum potential elastic strength which is achieved by the specimen under test.

### Energy to fracture

The area under a tensile curve is often used to calculate the energy required to fracture a specimen. A similar quantity can be determined from the bend-test curve, although in view of the particular method used to determine extension it is difficult to give an exact correlation between area and energy. Nevertheless, an approximate conversion can be used to compare specimens and it appears obvious that the greater the area under the curve the 'tougher' will be a specimen. In computing the values quoted in this paper corrected bend-test curves have been used.

### Effect of other properties on energy to fracture

Examination of the form of various selected bend-test curves enables an analysis to be made of the factors governing the area under the curve. Figure 7 shows a schematic curve, reduced to straight lines for the purpose of calculation.

The curve is divided into three portions:

OB = elastic region
BC = partial yield at constant maximum fibre stress
CD = continued yield with increasing fibre stress.

The area $A$ can be computed as follows:

$$A = \frac{L \cdot \lambda_L}{2} + L(\lambda_F - \lambda_L) + \frac{(\phi - L)(\lambda_F - \lambda_Y)}{2}$$

The factors which promote a high value of $A$ are, assuming that the breaking stress $\phi$ is constant; (*a*) a high value of limit of proportionality $L$, (*b*) a high value of plastic deformation, (*c*) a low value of the length BC.

The maximum possible area for a given value of breaking stress $\phi$ and a given total deformation is given by the curve OJE, i.e. when $L = \phi$. It is suggested, therefore, that the ratio $L/\phi$ may be used to express in part the extent to which the maximum area under the curve is attained.

The area under the curve is also related to the plastic deformation ($\lambda_F$–$\lambda_L$ in Fig. 7) and, as observed in Appendix III, these two values are directly involved in the conversion of the nominal breaking stress $B$ to 'true' breaking stress $\phi$. Also the ratio $\lambda_L/\lambda_F$ is a direct measure of the proportion of the specimen which has not yielded, and from this the value $Y$ of the percentage fibres yielded is readily

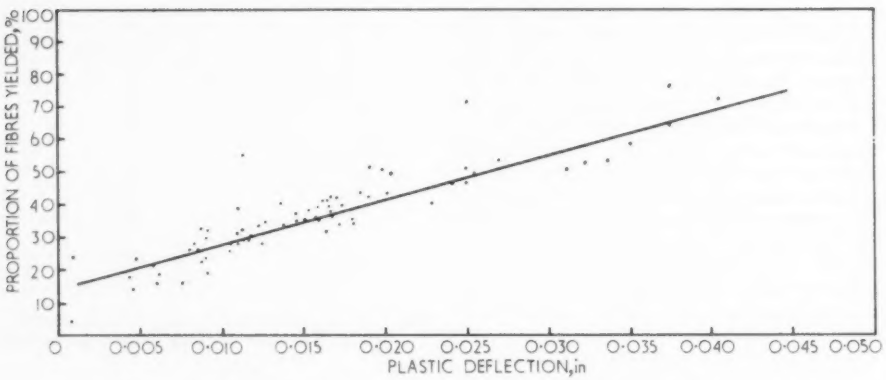



**Fig. 8—Correlation between *Y* and plastic deformation**

calculable. It is therefore suggested that the value of $Y$ is used as an appraisal of the plastic behaviour of the specimens: for brittle materials its value is 0 and for ideally ductile materials its value would be 100. The use of this, rather than the actual measurement of plastic deflection ($\lambda_F - \lambda_L$), is preferred on the grounds that it is independent of slight variations in specimen size, and also that it appears to be a more definite property of the material, as opposed to a somewhat arbitrary linear value.

The relationship between plastic deformation and $Y$ appears to be linear, as shown in Fig. 8 for a range of steels.

An empirical relationship is proposed to include the three factors considered, which is believed to represent a relative ' toughness ' or ' quality ' of the specimen in terms of the ultimate ideal, which may be defined as ' a specimen which gives the maximum elastic strength consistent with hardness and requires infinite energy to fracture.'

The ' quality factor ' $Q$ is defined as:

$$Q = \frac{L}{\theta} \times \frac{L}{\phi} \times Y$$

For any particular steel, $Q$ appears to be proportional to the energy to fracture $U$. Graphs correlating the two functions are shown in Fig. 9 for the two types of high-speed steel referred to in Tables IV and V.

### Modulus *E*

Measurements of the modulus of elasticity made during tests have shown rather large and unexpected variation (between $30 \times 10^6$ and $40 \times 10^6$ lb/in$^2$ although the majority of results fell in the range 33–36 $\times 10^6$ lb/in$^2$). The authors are of the opinion that, for high-speed steels, the origin of these variations may be found in the segregation of carbide, to which this type of steel is particularly prone.[28] One batch of specimens, prepared in such a way that one face of each was of unsegregated material and the other contained carbide stringers, gave consistently higher modulus figures when the segregated face was in tension than when the good face was in tension.

### General

It is presupposed in all the above discussions that hardness must be taken as a primary parameter, for the arguments break down if $H$ is below the acceptable level for the particular application. Most of the samples investigated by this technique have been of the order of 800 VPN or greater, and, as will be seen later, results become more erratic as the hardness level falls below about 750 VPN.

With this reservation, it is submitted that the methods outlined afford a sensitive and comparatively simple means of comparing samples of hardened tool steel, with respect to toughness or quality; these terms must still remain undefined but their significance may be better understood by consideration of the arguments advanced.

## EXPERIMENTAL WORK AND RESULTS

During the development of the test, a large number of experiments have been undertaken both to investigate the characteristic behaviour under test of materials whose properties were known, and to explore the properties where these were not known. A limited selection of these results is given below. In all

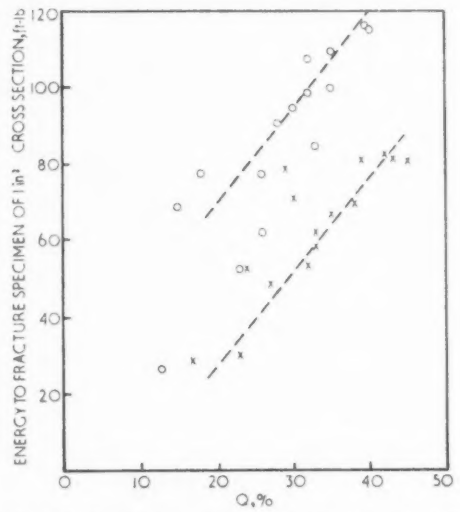



**Fig. 9—Relation between energy to fracture and *Q* for two steels**

**TABLE I**

**Effect of pre-hardening treatment on carbon tool steels (0·98%C–0·30%Si–0·24%Mn)**

| Ref. | Sample remarks | Hardness HD/30 | Limit of proportionality $L$, tons/in² | Nominal breaking stress $B$, tons/in² | Corrected breaking stress $\phi$, tons/in² | % Fibres yielded at fracture $Y$ | $\frac{L}{\theta}$ | $\frac{L}{\phi}$ | $Q$ |
|---|---|---|---|---|---|---|---|---|---|
| *A* | Unsatisfactory standard specimen | 693 | 154 | 170 | 151 | 13 | 1·35† | 1·02 | 18 |
| *B* | Satisfactory standard specimen | 754 | 154 | 206 | 162 | 34 | 1·13 | 0·95 | 36 |
| 9 | Hardened from as-forged condition | 754 | 135 | 135 | 135 | 0 | 1·00 | 1·00 | 0 |
| 10 | Hardened from as-forged condition | 760 | 150 | 150 | 150 | 0 | 1·09 | 1·09 | 0 |
| 13 | 'Normalized' from 800° C before hardening | 760 | 164 | 220* | 171* | 34* | 1·21 | 0·96 | 39 |
| 14 | 'Normalized' from 800° C before hardening | 765 | 161 | 211 | 167 | 31 | 1·14 | 0·96 | 34 |
| 17 | Annealed 780° C before hardening | 751 | 154 | 220* | 167* | 39* | 1·14 | 0·92 | 41 |
| 18 | Annealed 780° C before hardening | 763 | 156 | 178 | 152 | 18 | 1·12 | 1·02 | 21 |

* Specimen unbroken at limit of machine. Recorded value is highest observed reading
† Value subject to error due to low hardness, below limit of experimental curve for $\theta$

cases the results have been abbreviated. It is normal practice to test batches of 10 specimens with identical treatment, and the arithmetic means of the various values have been quoted.

### Effect of pre-hardening treatment on carbon tool steel

Two samples, *A* and *B*, of 1·2%C tool steel were examined, *A* being reported to have shown inferior service to *B*. Samples of similar steel were tested after the treatments as given in Table I. In all cases the hardening and tempering treatments were identical.

The results clearly demonstrate the inferior properties of samples 9 and 10 (hardened from the as-forged condition) compared with those of samples annealed or 'normalized' before hardening. The low hardness of sample *A*, however, gives a rather misleading result; this emphasizes the importance of ensuring correct hardness before applying the test. In this particular case, the low value of $Y$ demonstrates the inferior quality.

The reason for the difference in behaviour between samples *A* and *B* was sought by metallographic examination. Figures 10 and 11 show that the brittleness of *A* apparently arose from the presence of undesirable carbide networks which were absent in *B*. Similar observations were made on the remaining samples, those hardened without prior treatment showing grain boundary carbide and the remainder evenly dispersed carbide. It may be claimed therefore that the bend test gave, in this instance, quantitative confirmation of the metallographic observation, which must always be the result of a personal assessment.

### Effect of tempering temperature on properties of En 31 steel

Samples of En 31 steel (1%C–1%Cr) were oil-quenched, after suitable prior annealing, from 820° C and tested in the respective conditions indicated in Table II. The results are in accordance with expectation, but serve to demonstrate the following points:

(i) The value of $L/\phi$ rises with tempering temperature, confirming the findings of Hamaker *et al*[21] and supporting the earlier arguments
(ii) The embrittlement after tempering at 250° C is prominently revealed by the low values of $Y$ and $Q$ for this series

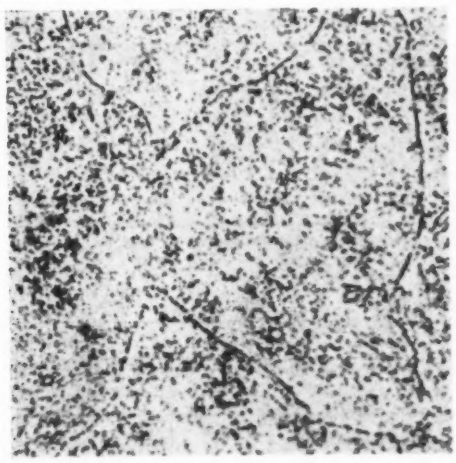


Fig. 10—Sample A, Table I × 1 000

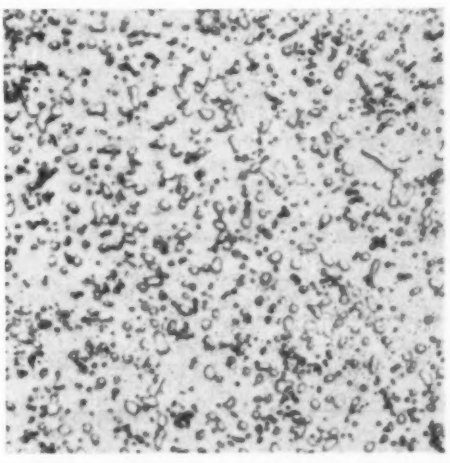


Fig. 11—Sample B, Table I × 1 000

Both specimens etched in 5% picral

**TABLE II**

**Effect of temperature on properties of En 31 steel (0·96%C–0·24%Si–0·73%Mn–1·35%Cr–0·12%Ni)**

| Treatment | Hardness HD/30 | Limit of proportionality $L$, tons/in² | Nominal breaking stress $B$, tons/in² | Corrected breaking stress $\phi$, tons/in² | % Fibres yielded at fracture $Y$ | $\frac{L}{\theta}$ | $\frac{L}{\phi}$ | $Q$ | Plastic deformation, in. | Un-notched impact value, ft-lb |
|---|---|---|---|---|---|---|---|---|---|---|
| As-quenched | 865 | 94 | 104 | * | * | * | * | * | 0·0014 | 11 |
| Tempered 100° C | 836 | 88 | 149 | 104 | 44 | 0·60 | 0·84 | 22 | 0·0082 | 11 |
| Tempered 150° C | 768 | 138 | 213 | 152 | 44 | 0·90 | 0·91 | 36 | 0·0191 | 16 |
| Tempered 200° C | 722 | 119 | 223 | 140 | 48 | 1·00 | 0·85 | 41 | 0·0227 | 38 |
| Tempered 250° C | 681 | 146 | 187 | 146 | 17 | 1·14 | 1·00 | 19 | 0·0139 | 30 |
| Tempered 300° C | 619 | 148 | 224 | 172 | 46 | † | 0·86 | approx 40 | 0·0246 | 45 |

* Results not calculated owing to excessive scatter
‡ Not calculated because of low hardness

(iii) The values of $Y$ and $Q$ vary with tempering temperature in the same way as the un-notched Izod properties given in Table II for samples (0·45 in. dia.) prepared from bars given the same heat treatment as the bend-test specimens

(iv) The optimum combination of hardness, strength, and ductility is attained by tempering at about 200° C, which agrees well with the commercial practice specified for this type of steel.

**Comparison of three types of high-speed steel**

The purpose of this series of tests was to determine the extent to which the bend test is capable of discriminating between (*a*) the steels themselves and (*b*) similar steels of slightly differing properties, imparted by small variations in austenitizing temperature. Heat-treatment details and results are given in Table III, and a brief analysis of the results is discussed below.

18–4–1 *steel*—The value of $L$ does not appear to change with austenitizing temperature, but the lower hardness of series 349 specimens means that the $L/\theta$ ratio is higher in this case, with a corresponding enhancement of $Q$, which is further increased by the higher $Y$ value. Also in the same series, the value of $L/\phi$ is slightly higher than in series 348. These observations confirm the common view that increased toughness is achieved by a reduction in austenitizing temperature, as in the heat treatment of complex tools such as taps and dies.

18–4–2 + 10%Co *steel*—All specimens in series 350 and 351 failed without plastic deformation, hence zero values are obtained for $Y$ and $Q$, and the value of $L/\phi$ becomes unity. It is, however, shown that a decrease in austenitizing temperature is accompanied by an increase in breaking stress, and hence of

**TABLE III(*a*)**

**Comparison of three types of high-speed steel**

| Sample series | Type | Austenitizing temp., °C | Hardness HD/30 | Limit of proportionality $L$, tons/in² | Nominal breaking stress $B$, tons/in² | Corrected breaking stress, $\phi$ tons/in² | % Fibres yielded at fracture $Y$ | $\frac{L}{\theta}$ | $\frac{L}{\phi}$ | $Q$ |
|---|---|---|---|---|---|---|---|---|---|---|
| 348 | 18–4–1 | 1 290 | 854 | 163 | 213 | 176 | 25 | 0·94 | 0·93 | 22 |
| 349 | 18–4–1 | 1 270 | 806 | 165 | 224 | 171 | 34 | 1·06 | 0·96 | 35 |
| 350 | 18–4–2 + 10%Co | 1 340 | 882 | 123 | 123 | 123 | 0 | 0·67 | 1·00 | 0 |
| 351 | 18–4–2 + 10%Co | 1 320 | 884 | 178 | 178 | 178 | 0 | 0·95 | 1·00 | 0 |
| 352 | 6–5–2 | 1 240 | 781 | 148 | 250* | 186* | 56* | 1·03 | 0·80 | 46 |
| 353 | 6–5–2 | 1 220 | 775 | 134 | 250* | 169* | 61* | 0·98 | 0·79 | 47 |

* Specimen not fractured at limit of machine. Values recorded are highest observed

**TABLE III(*b*)**

**Composition of high-speed steels**

| Type | Element, % | | | | | | | |
|---|---|---|---|---|---|---|---|---|
| | C | Si | Mn | W | Cr | V | Co | Mo |
| 18–4–1 | 0·73 | 0·57 | 0·31 | 18·2 | 4·13 | 1·07 | 0·26 | 0·4 |
| 18–4–2 + 10%Co | 0·79 | 0·29 | 0·22 | 18·65 | 4·47 | 1·85 | 9·2 | 0·95 |
| 6–5–2 | 0·80 | 0·30 | 0·27 | 6·5 | 4·03 | 1·8 | ... | 4·94 |

**TABLE IV**
**Effect of heat-treatment on properties of 6–5–2 steel**

| Series | Quench temperature, °C | Tempering temperature, °C | Hardness HD/30 | Limit of proportionality $L$, tons/in² | Nominal breaking stress $B$, tons/in² | Corrected breaking stress $\phi$, tons/in² | % Fibres yielded at fracture $Y$ | $\frac{L}{\theta}$ | $\frac{L}{\phi}$ | $Q$ | Modulus $E$, lb/in² × 10⁶ | Energy*/in³ $U$, ft-lb | Plastic deflection, in. |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| A | 1 260 | 550 | 886 | 142 | 214 | 157 | 37 | 0·76 | 0·92 | 26 | 33·0 | 48 | 0·0146 |
| B | 1 200 | 550 | 775 | 104 | 228 | 158 | 70 | 0·72 | 0·66 | 33 | 34·4 | 78 | 0·0411 |
| C | 1 230 | 570 | 835 | 140 | 238 | 166 | 54 | 0·85 | 0·84 | 39 | 33·3 | 75 | 0·0316 |
| D | 1 230 | 550 | 838 | 142 | 233 | 166 | 51 | 0·85 | 0·85 | 38 | 35·1 | 69 | 0·0301 |
| E | 1 230 | 530 | 825 | 99 | 200 | 141 | 60 | 0·64 | 0·71 | 26 | 36·4 | 47 | 0·0215 |

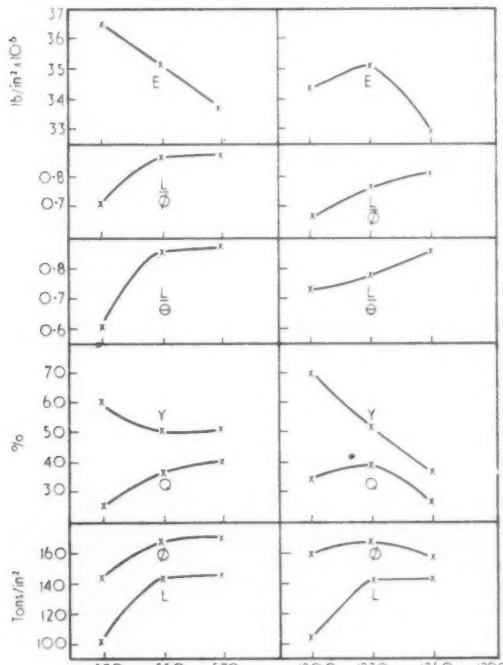
* Computed for 1 in. square cross section; individual samples varied in dimensions to a sufficient extent to make direct comparison of doubtful value

energy to fracture, once again in accordance with commercial heat-treatment principle for this type of steel.

6–5–4–2 *high-speed steel*—The hardness level of these specimens was such that they were in a condition unacceptable for cutting tools. It is possible to discriminate between the two series 352 and 353 by reference to the quantities $L$ and $\phi$, otherwise it appears that both treatments have given the optimum properties attainable consistent with the low hardness.

**Effect of heat treatment on 6–5–4–2 steel**

A series of specimens of 6–5–4–2 high-speed steel (not the same batch as in the previous section) was examined in the conditions shown in Table IV. The heat-treatment variations covered a range of tempering temperature with constant austenitizing temperature, and a range of austenitizing temperature with constant tempering temperature. The variation of properties is shown in Fig. 12, from which it is possible to deduce the optimum heat treatment for any given properties.




$E$ = Modulus of elasticity
$L$ = Limit of proportionality
$\theta$ = Optimum value of $L$ (for a given hardness)
$\phi$ = 'True' breaking stress
$Y$ = % of total specimen yielded at fracture
$Q$ = Quality factor

**Fig. 12—Effect of heat treatment on 6–5–2 steel**

It is of interest to note the linear decrease in modulus $E$ with rising tempering temperature but, as stated earlier, the exact significance of this is not understood.

**Determination of optimum heat treatment for a special steel**

Samples of an experimental high-speed steel were submitted. One series, of unknown heat treatment, were reported to give excellent service, and the four remaining series had been heat treated by various thermal cycles in an attempt to reproduce the properties of the standard. Results are given in Table V, from which it is apparent that treatments 1 and 4 are superior to 2 and 3 but still give properties inferior to the standard. It is probable in this case that a further improvement could have been obtained by a slight increase in tempering temperature.

## SUMMARY AND CONCLUSIONS

(1) A modified bend-test apparatus is described, which is specifically designed for the testing of hardened tool steels at room temperature or elevated temperature.

(2) An analysis of the bend test has been made, and a method developed for approximate correction of the bending laws to enable a more accurate calculation of stresses in a specimen subject to plastic deformation.

(3) The findings of Hamaker *et al*[21] have been confirmed, in that limit of proportionality $L$ appears to be related to hardness provided the specimen is adequately tempered. From this relationship a value $\theta$ is found for the 'potential' limit of proportionality corresponding to a given hardness and the ratio $L/\theta$ taken as a measure of the effectiveness of tempering of a specimen under test.

(4) It has been assumed that the energy to fracture is represented by the area under the corrected bend-test curve, and for a given breaking stress $\phi$ the following factors contribute to an increase in this area:

(i) A high limit of proportionality $L$ compared with breaking stress $\phi$
(ii) A high value of plastic deformation.

**TABLE V**

**Effect of varying treatment on special experimental high-speed steel**

| Treatment no. | Hardness HD/30 | Limit of proportionality $L$, tons/in² | Nominal breaking stress $B$, tons/in² | Corrected breaking stress $\phi$, tons/in² | % Fibres yielded at fracture, $Y$ | $\frac{L}{\theta}$ | $\frac{L}{\phi}$ | $Q$ | Plastic deformation, in. |
|---|---|---|---|---|---|---|---|---|---|
| Reference Standard | 851 | 179 | 234 | 180 | 41 | 1·04 | 0·99 | 42 | 0·0183 |
| 1 | 878 | 151 | 226 | 170 | 41 | 0·83 | 0·89 | 30 | 0·0182 |
| 2 | 894 | 144 | 171 | 149 | 16 | 0·77 | 0·97 | 11 | 0·0058 |
| 3 | 918 | 150 | 172 | 152 | 15 | 0·76 | 0·98 | 10 | 0·0051 |
| 4 | 905 | 164 | 253 | 179 | 34 | 0·85 | 0·92 | 27 | 0·0134 |

(5) In calculating the corrected breaking stress $\phi$, a value $Y$ is obtained which represents the proportion of the specimen stressed beyond the yield point at fracture. This value has been shown to be proportional to the linear measurement obtained for plastic deformation, and is used in calculations under 4(ii) instead of the linear value.

(6) An empirical formula giving an overall assessment of the toughness of a sample is suggested, embracing the three factors governing the area under the bend-test curve. The value $Q = (L/\theta)(L/\phi)Y$ is shown to be proportional to the energy $U$ obtained from the bend-test curve, but the slope of the line appears to vary from steel to steel.

(7) By means of the bend test, using the methods of calculation described, it is possible to discriminate between the properties of a range of tool steels of equal hardness, in terms of elastic behaviour, plastic behaviour, and energy required to fracture, but not, as yet, in terms of cutting ability or wear.

(8) The importance is stressed of ensuring that the hardness of a steel is adequate for the particular application before any attempt at appraisal of its properties in terms of toughness is made.

(9) The principles discussed have been demonstrated by the results from selected *ad hoc* investigations.

#### Acknowledgments

This investigation was initiated by Dr M. L. Becker, formerly Head of the Metallurgy Division, to whom the authors are indebted for helpful advice and encouragement. Thanks are also due to members of the Tool Steels Committee for their advice and cooperation in supplying and heat treating samples used in the investigation, to the design and machine shop staff for their help in developing the apparatus, to various colleagues who have, from time to time, assisted with the experimental work, and to Mr J. Clayton-Cave for helpful discussion.

## APPENDIX I

### Calculation of stresses using simple bending theory

Classical bending theory may be applied to the results obtained using the apparatus described, as follows. If a load $W$ is applied to the specimen whose breadth and depth are $b$ and $d$ respectively, the bending moment over the central span can be computed from the diagram in Fig. 13: $M = W/4$.

Using the formula $F/y = M/I = E/R$, and taking $y = d/2$, a value is obtained for maximum fibre stress

$$F = \frac{My}{I} = \frac{W}{4} \cdot \frac{d}{2} \cdot \frac{12}{bd^3} = \frac{3W}{2bd^2} = W\sigma$$

where $\sigma = 3/2bd^2$.

From calibration, it is known that $W = k_1 G$ where $G$ is the galvanometer reading and $k_1$ an experimental constant, $\therefore F = k_1 G\sigma$.

Assuming circular arc bending over the central span, the value of $\Delta$, the deflection at the mid-point, gives the radius of curvature $R$ from $R = \frac{1}{8\Delta}$ (Fig. 14).

The value of $\Delta$ is obtained from the dial reading $\lambda$:

$$\Delta = k_2\lambda,$$

where $k_2$ is an experimental constant.

By substitution in the formula $F/y = E/R$, the modulus $E$ can be found.

## APPENDIX II

### Modification to allow for plastic deformation

In the classical theory, used in Appendix I, one of the basic assumptions is that specimens behave in an ideally elastic manner throughout the test. This condition is violated immediately the limit of proportionality has been exceeded, and it is necessary to examine the correction factor which must be applied to enable the accurate calculation of stresses in a deformed specimen.

Figure 4*a* shows the stress distribution in a specimen subject only to elastic strain. If the breadth and depth of specimen are $b$ and $d$ respectively then the stress on an element, depth $dy$, distance $y$ from the neutral axis, is given by $F = 2\phi y/d$ where $\phi$ is the maximum stress (i.e. when $y = d/2$).

The area of the element is $b \cdot dy$, and the force on the element is therefore $2\phi y \cdot b \cdot dy/d$, and the moment of this force about the neutral axis is $M = \frac{2\phi b\, y^2 dy}{d}$

Over the whole section, the moment of resistance $MR$ is given by:

$$MR = \int_{d/2}^{d/2} \frac{2b\phi}{d} \cdot y^2\, dy = \frac{\phi b d^2}{6}$$

which is the result obtained by classical theory.

Now consider a stress distribution as in Fig. 4*b*. It is here assumed that the outer fibres have yielded and are at constant stress $\phi$, and that over the unyielded central portion of a specimen classical theory still holds.

Consider an element, depth $dy$, distance $y$ from the N.A., within the yielded area.

Stress on element $= \phi$
Area of element $= b \cdot dy$
Force on element $= \phi b \cdot dy$
Moment about N.A. $= \phi b \cdot y \cdot dy$

The total moment of resistance, taking into consideration the two yielded portions, is therefore,

$$M_1 = 2\int_a^{d/2} \phi by \cdot dy = \phi b\left[\frac{d^2}{4} - a^2\right]$$

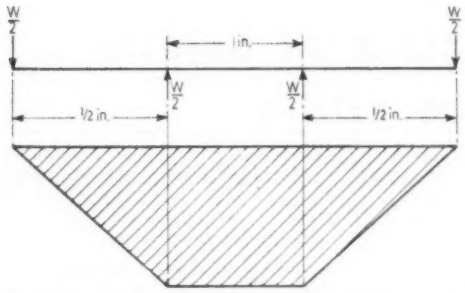



**Fig. 13—Distribution of force and bending moment diagram, not to scale**

Applying the normal theory to the unyielded portion, the moment of resistance is: $M_2 = \phi b \cdot \frac{2a^2}{3}$

Over the whole specimen:

$$M = M_1 + M_2 = \frac{\phi b}{12}[3d^2 - 4a^2] = \phi Z^1$$

where $Z^1$ = modulus of section.

For the unyielded specimen $Z = bd^2/6$.

In calculating stresses, $Z$ has formerly been used. It is now suggested that the value $Z^1$ would be more accurate for use in the partially yielded condition. If the stress $F$ given by $F = M/Z$ is compared with the value of $\phi$ given by $\phi = M/Z^1$, then:

$$\frac{F}{\phi} = \frac{Z^1}{Z}$$

$$\phi = F \cdot \left(\frac{2d^2}{3d^2 - 4a^2}\right) = F \cdot \left[1 \cdot 5 - \frac{a^2}{2\left(\frac{d}{2}\right)^2}\right]^{-1}$$

It is appreciated that this relationship would be altered if the neutral axis ceased to pass through the centre of the specimen. The same objection applies, however, to calculations made by the normal classical method, and it is submitted that the revised method gives a nearer approach to the true value.

## APPENDIX III

### Application of the result of Appendix II to bend test results

It is apparent from Fig. 4*b* that the stress at a point distance $a$ from the neutral axis is equal to $L$, the stress at the limit of proportionality. It is assumed that the unyielded portion of the specimen can be treated apart from the yielded portion, as follows:

Applying the standard formula to this portion alone, $F/y = E/R$, where $R = 1/8\Delta$, (*see* Appendix I)

$L/a = E \times 8\Delta$ or $a = L/8E\Delta$.

For the whole specimen, assuming a condition as in Fig. 4*a* where the maximum fibre stress is $L$, and the value of $\Delta$ is given by $\Delta_L$,

$$\frac{L}{d/2} = E \times 8\Delta_L$$

$$\frac{d}{2} = \frac{L}{8E\Delta_L}$$

$$\frac{a}{d/2} = \frac{L}{8E\Delta} \times \frac{8E\Delta_L}{L} = \frac{\Delta_L}{\Delta} = \frac{\lambda_L}{\lambda}$$

$$\therefore a = \frac{d}{2} \cdot \frac{\lambda_L}{\lambda}$$

where $\lambda_L$ and $\lambda$ are dial gauge readings.

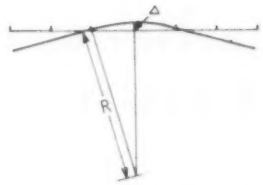



**Fig. 14—Calculation of radius of curvature**

Calculation is simplified by writing $\eta = 2a$, where $\eta$ represents the total thickness of unyielded specimen.

$$\eta = \frac{d \cdot \lambda_L}{\lambda}$$

From Appendix II:

$$\phi = F \cdot \left[1 \cdot 5 - \frac{a^2}{2 \cdot (d/2)^2}\right]$$

$$= 2F \cdot \left[3 - \frac{\eta^2}{d^2}\right]^{-1} = 2F \cdot \left[3 - \left(\frac{\lambda_L}{\lambda}\right)^2\right]^{-1}$$

The ratio $(1 - \eta/d) \times 100$ gives $Y$, the percentage fibres yielded.

### References


1. J. V. Emmons: *Proc. ASTM*, 1931, vol. 31, Pt. II, pp. 47–76.
2. J. V. Emmons: *Trans ASST*, 1932, vol. 19, pp. 289–318.
3. J. V. Emmons: *Ibid.*, 1933, vol. 21, pp. 193–221.
4. Yu. A. Geller: *Savodskaya Laboratoriya*, 1955, vol. 21, No. 5, pp. 594–601. (*ISI* Translation No. 526, 1957).
5. G. V. Luerssen and O. V. Greene: *Proc. ASTM*, 1933, vol. 33, Pt. II, pp. 315–327.
6. G. V. Luerssen and O. V. Greene: *Trans ASM*, 1934, vol. 22, 311–337.
7. G. V. Luerssen and O. V. Greene: *Ibid.*, 1935, vol. 23, pp. 861–878.
8. O. V. Greene and R. D. Stout: *Proc. ASTM*, 1939, vol. 39, pp. 1292–1298.
9. O. V. Greene and R. D. Stout: *Trans ASM*, 1940, vol. 28, pp. 277–298.
10. A. Knight: *ARE* Metallurgical Report. Ministry of Supply, Unpublished, 1950.
11. A. H. Grobe and G. A. Roberts: *Trans ASM*, 1950, vol. 42, pp. 686–707.
12. H. Carr: *Iron Steel*, 1950, vol. 23, pp. 383–388.
13. Carpenter Steel Co. Service Bulletin, vol. 2, No. 25.
14. W. E. Bancroft and W. W. Wight: *Met. Prog.*, 1958, vol. 53, pp. 545–555.
15. S. and T. Memo. 17/55. Ministry of Supply.
16. R. K. Barry: *Trans ASST*, 1927, vol. 12, pp. 630–634.
17. R. K. Barry: *Ibid.*, 1926, vol. 10, pp. 257–265.
18. A. H. d'Arcambal: *Ibid.*, 1922, vol. 2, pp. 586–601.
19. K. Amberg: *Jernkont. A.*, 1929, vol. 84, pp. 280–299.
20. A. H. Grobe and G. A. Roberts: *Trans ASM*, 1948, vol. 40, pp. 435–471.
21. J. C. Hamaker, V. C. Strang, and G. A. Roberts: *Ibid.*, 1957, vol. 49, pp. 550–573.
22. J. Clayton-Cave and E. Ineson: *BISRA* Report. MG/A/149/52. (Now superseded by Report, MG/L/60/57.)
23. R. S. Rose: *Met. Prog.*, 1940, vol. 37, pp. 407–411.
24. "Machining, Theory and Practice" p. 409, Cleveland, Ohio, 1950: American Society for Metals.
25. C. Wainwright and L. W. Nickols: *Metallurgia*, 1950, vol. 42, pp. 401–404.
26. R. K. Roark: "Formulas for Stress and Strain," 2nd edn., pp. 111–114, 1943, London and New York, McGraw Hill.
27. S. R. Williams: "Hardness and Hardness Measurement," pp. 42–48, 1942, Cleveland, Ohio.
28. E. Ineson and G. Hoyle: *Met. Treatment*, 1956, vol. 23, pp. 257–262.

# The kinetics of the formation of bainite in high-purity iron–carbon alloys

*By S. V. Radcliffe, Ph.D., A.I.M., and E. C. Rollason, M.Sc., Ph.D.*

SYNOPSIS

The isothermal transformation of austenite to bainite at temperatures in the range from 200–450° C has been studied in three iron–carbon alloys (0·77, 1·07, and 1·20 wt-% C) and a plain carbon steel (0·65 wt-% C) by the measurement of changes in electrical resistivity. The processes of austenitizing and rapidly cooling wire specimens to the reaction temperature were accomplished by direct electrical heating followed by gas-quenching. The resistivity and temperature of the specimen during quenching and subsequent isothermal reaction were recorded continuously. It has been established that a marked change in the kinetics of the transformation occurs over the temperature range 300–350° C. In addition, the formation of bainite below this range is associated with changes in resistivity and specific volume which indicate that the ferrite in the structure becomes increasingly supersaturated with carbon as the reaction temperature is lowered. No accompanying tetragonality of the ferrite was discernable from X-ray diffraction measurements on fully reacted specimens. The measured kinetics show that the formation of bainite is not a simple, growth-controlled reaction and suggest that it is initiated by the nucleation of ferrite at a rate dependent on the temperature and on the composition of the austenite. 1576

## INTRODUCTION

THE TRANSFORMATION OF AUSTENITE to the acicular aggregate of ferrite and carbide known generally as bainite occurs in both plain carbon and alloy steels over a temperature range intermediate between those of the pearlite and martensite reactions. Bainite may be made to form isothermally by quenching from the austenite region to a temperature within this intermediate range, or athermally by cooling through it. Despite the interest shown in the reaction since it was first delineated in the late 1920's there is still insufficient experimental data to develop a satisfactory model for the mechanism of this mode of austenite decomposition. Previous studies of bainite have been reviewed recently,[1,2] and it is sufficient here to summarize the established characteristics and suggested models of the transformation.

The work of Ko and Cottrell,[6] and Tsuya,[7] has established that the reaction is one of nucleation and coherent growth of bainite in the austenite. The orientation relationships and metallographic observations suggest that coherent ferrite is the nucleating phase, but whether the ferrite is carbon-free or supersaturated with carbon has only been ascertained indirectly. Ko's conclusion that the characteristics of bainite gradually change as the reaction temperature is lowered does not agree with the evidence provided by published electron-micrographs.[3] Examination of these micrographs indicated that the change from the 'longitudinal carbide-stringer' structure of upper bainite to the 'transverse carbide-platelet' structure of lower bainite occurs over a fairly narrow range of temperature in the region of 310–350° C. The change in microstructure suggests that the kinetics of the bainite reaction should differ above and below this temperature range, but no such difference has yet been established.

For plain carbon steels, the existence of a sharp upper limit $B_s$ to the bainite temperature range, which is characteristic of alloy steels, has not been established. Similarly, the dependence on temperature of the complete transformation to bainite $B_f$ which has been found for many alloy steels is not a proven characteristic of plain carbon steels. An accurate determination of the kinetics of the transformation of austenite at temperatures where overlap of the pearlite and bainite ranges occurs is needed to ascertain whether both reactions take place simultaneously, or whether bainite formation ceases before the onset of pearlite.

Although the distinctive features of bainite make the reaction important from the point of view of the development of a satisfactory theory of phase transformations, knowledge of the mechanism of its formation is lacking. This is sharply revealed by the several different and often contradictory suggestions based on theoretical analyses which have been put forward to explain the reaction or particular aspects of it.[5, 26, 27, 28] A principal difficulty in developing a satisfactory hypothesis is that most experimental studies have been carried out on alloy steels. Whilst the addition of alloying elements is helpful in isolating the bainite reaction by lowering the $M_s$ temperature, reducing the reaction rate, and separating the bainite and pearlite ranges, their presence must also complicate the mechanism of the basic reaction in binary iron–carbon alloys.

Two principal mechanisms have been suggested to account for the characteristics of the reaction. In one, bainite is considered to form initially as supersaturated ferrite by a lattice-shearing process, and to grow at a rate controlled by the diffusion of carbon into the surrounding austenite or to carbide particles precipitated within the ferrite regions. In the other, bainite is considered to form directly from austenite as an aggregate of ferrite and carbide. In both cases, ferrite is held to be the nucleating phase. The observations of the reaction made by hot-stage microscopy and the majority of those made by electron-microscopy suggest the general acceptance of the

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diffusion-controlled shear process, although there is metallographic evidence for the direct process in a few alloy steels. Neither of these mechanisms has provided a model which will fully describe the kinetics of the reaction.

Recently, both types of mechanism have been combined by Krisement and Wever[10] in an interesting attempt to account for the effects of both temperature and carbon content on the general characteristics and the kinetics of the reaction. The analysis is based on the concept of 'double-precipitation', i.e. alternate precipitation of carbide and ferrite in the growth direction. Although this leads to an explanation of the characteristics of the reaction (with the exception of that of incomplete transformation in alloy steels), the agreement with experimental data is not good. The real weakness of the theory is that the 'double-precipitate' model does not appear to conform to the established microstructures of bainite. Thus, although the theory leads to the postulate of a decrease in 'activation energy' with decreasing carbon content, which is confirmed qualitatively by available experimental data for alloy steels, the agreement may only be fortuitous. In spite of these drawbacks the analysis is of considerable importance since any description of the mechanisms of bainite formation must attempt to account for the effects of both temperature and composition on the kinetics of the reactions.

The present investigation was undertaken to determine the effects of temperature and carbon content on the kinetics of the isothermal formation of bainite in high-purity iron–carbon alloys. The use of such alloys avoids any possible complications associated with the presence of alloying elements or impurities. In the experimental method adopted, the progress of isothermal transformation at temperatures in the range 200–450° C was followed by high-speed measurements of changes in the electrical resistivity of fine wire specimens containing 0·65–1·2 wt-%C. These measurements were supplemented by examination of the structures of fully reacted specimens by light microscopy and X-ray diffraction at room temperature.

## EXPERIMENTAL METHODS

### Preparation of alloys

The iron–carbon alloys used were prepared directly by gas-curburizing specimens of pure iron in the form of wires 0·010 in. dia. and 8 in. long. The wires were made from a vacuum-melted, deoxidized iron (99·96% Fe) supplied by the British Iron and Steel Research Association. Carburizing was carried out at a constant temperature of 925° C in an atmosphere of nitrogen saturated with benzene vapour. The range of carbon contents, 0·77, 1·07, and 1·2%, was obtained by varying the carburizing time. The carbon contents were determined from electrical resistivity measurements according to the simple relationship recently established.[11]

The preparation of the alloys in this manner was preferred to a melting technique in order to minimize the introduction of impurities and avoid the difficulties of fabricating small, homogeneous specimens of high-carbon alloys from cast ingots.

The commercial plain carbon steel (0·65%C–0·35%Mn) was obtained in the form of 0·095-in. dia. wire.

### Transformation and recording techniques

The most accurate determination of the kinetics of an isothermal transformation is afforded by the continuous measurement at the reaction temperature of the accompanying changes in physical properties. Electrical resistivity, which has already been used for transformation studies in steel was selected as the most suitable property for the study of bainite formation.[12, 15]

The technique developed to effect and follow the transformation uses a wire specimen supported in an evacuated furnace tube which is controlled to the desired temperature of isothermal reaction. The specimen is heated independently to the austenitizing temperature by the direct passage of an electric current. On completion of austenitizing, the specimen is quenched back to the reaction temperature with a blast of inert gas after switching off the heating current. Simultaneously, a small direct current is introduced and the temperature of the wire and the potential drop along its length are recorded. The electrical resistivity of the specimen is determined by a comparison with the potential drop across a known resistance which is included in the same circuit.

To reduce heat conduction from the specimen, the thermocouple and potential wires, which are spot-welded to it, are only 0·002 in. dia. The austenitizing treatment was carried out *in vacuo*. To prevent oxidation and decarburization, and to ensure a constant temperature along the length of the specimen by minimizing the heat loss due to convection, the quenching gas, purified nitrogen, was admitted to the system by means of a rapid-action hand-operated valve.

From the instant of quenching, the potential drop along the specimen and the output of the specimen thermocouple were recorded on two Honeywell-Brown 'Electronik' strip-chart potentiometers. These instruments have a high speed of response to changes in input signals, a full chart-width deflection of 11 in. (50 mV) in 1·2 s, and a high maximum chart speed of 8·0 in./min.

For the study of more rapid changes a technique was developed of recording on a Cossor double-beam cathode-ray oscillograph.

## EXPERIMENTAL RESULTS

### Resistivity measurements

Sets of isothermal reactions at temperature intervals of about 25° C were carried out with specimens of each of the four materials. All specimens were austenitized under identical conditions; 3 min at 1 100° C. The resulting austenite grain size was ASTM3; equivalent to an average of 4·5 grains on the wire cross-section.

Initially, measurements of the change in the potential drop along the specimen as the reaction proceeded were plotted as a function of time from the instant of quenching. In all the reactions observed the change in potential took place smoothly with time. For a direct comparison of transformations carried out at different temperatures, graphs of the total percentage

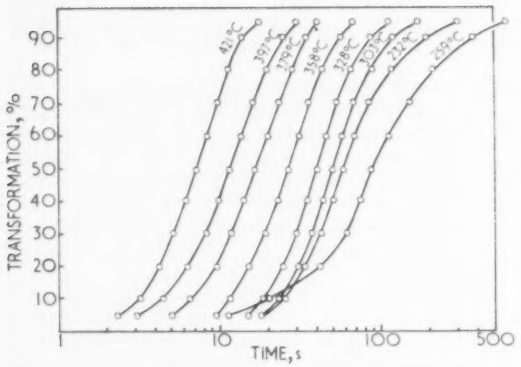


**Fig. 1—Isothermal reaction curves for plain-carbon steel (0·65%C)**

change in specimen potential plotted as a function of time were derived from the reaction data. An illustrative set of curves obtained in this way for the plain-carbon steel is shown in Fig. 1.

Metallographic and X-ray diffraction studies of the fully reacted specimens established that no untransformed austenite remained on completion of the changes in resistivity. In view of this and of the correlations obtained by previous investigators between the progress of transformation and the accompanying changes in resistivity, the ordinate in Fig. 1 has been taken to represent the percentage of transformation of austenite to bainite.

In Figures 2 and 3 the times taken at the various reaction temperatures to attain 0, 5, 50, and 95% transformation are shown for two alloys in the form of isothermal transformation diagrams. At the lower temperatures a marked acceleration of the start of reaction is apparent.

Absolute measurements of resistivity were not required for the derivation of the reaction curves. However, to compare the different conditions of a specimen during its transformation cycle, such measurements were made (*a*) at room temperature after insertion in the apparatus, (*b*) at the reaction temperature before austenitizing, (*c*) at the reaction temperature after austenitizing, but before the reaction began, and (*d*) at the reaction temperature on completion of the transformation. The resulting data for the plain carbon steel and for the 0·77%C alloy are presented in Figs. 4 and 5 in the form of graphs of the change in resistivity with temperature of three different structures, pearlite, bainite, and austenite. The set of three points corresponding to these structures at each reaction temperature was obtained from a single transformation cycle on one specimen. For reaction temperatures below $M_s$, the indicated 'austenite' and 'bainite' resistivities represent those of mixtures of austenite with martensite, and bainite with tempered martensite.

The results for the 1·07% and 1·2%C alloys are similar to those for the 0·77%C alloy. In all the materials examined, the resistivity of austenite decreases discontinuously as the temperature is lowered. Further, the resistivity of a specimen reacted fully to bainite does not correspond to that

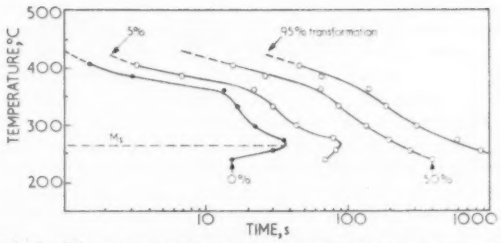


**Fig. 2—Time/temperature transformation diagram for iron–carbon alloy (0·77%C)**

of the same specimen in its original pearlitic condition at all temperatures. In the case of the plain carbon steel, the resistivity in the bainite condition is always higher than in the pearlitic condition at the same temperature.

In contrast, for all three pure alloys the resistivity of a specimen reacted to bainite at temperatures above some 350° C coincides with that of the specimen in its original pearlitic condition at the same temperature. Below this temperature the resistivity of a specimen reacted to bainite is always greater than that of the corresponding pearlitic condition. The difference increases as the reaction temperature is lowered, as is shown also by the curves for the plain carbon steel.

The failure of the resistivity to fall to the value expected for a ferrite–carbide aggregate of equilibrium composition was accompanied by dilational changes. On isothermal reaction within the temperature ranges over which the resistivities of the pearlitic and bainitic conditions diverge, specimens became slightly buckled, indicating an expansion beyond their original length, since the wires were brazed between rigid supports. Such differences in resistivity and dilation were not observed for specimens allowed to transform to pearlite after a similar austenitizing treatment.

#### Examination of reacted specimens by X-ray diffraction

In view of the anomalous changes in resistivity and dilatation noted during isothermal transformation, an X-ray study was made of specimens reacted fully to bainite, pearlite, and martensite. It has been suggested,[6, 7] that excessive dilational changes observed on the formation of bainite in alloy steels are associated with supersaturation of ferrite with carbon. The changes noted in the present study were of such a magnitude that it was anticipated that any associated changes in the lattice parameter of the

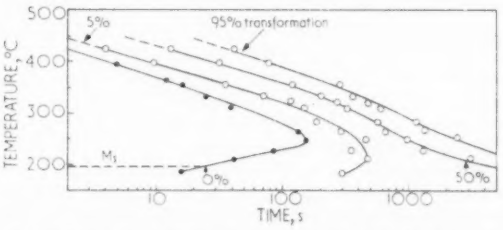


**Fig. 3—Time/temperature transformation diagram for iron–carbon alloy (1·07%C)**

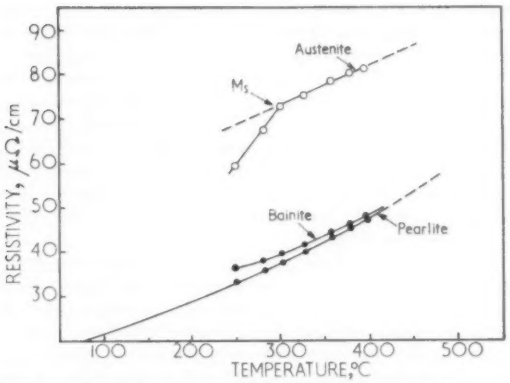


**Fig. 4—Changes in resistivity during isothermal reaction of plain carbon steel (0·65%C)**

ferrite due to retention of carbon might be detected by a corresponding shift in the position of the lines of the X-ray diffraction pattern.

Using iron K-alpha radiation, Debye-Scherrer patterns were obtained from steel wires (*a*) reacted to pearlite by slow cooling through the critical range, (*b*) reacted isothermally at different temperatures in the bainite range, and (*c*) quenched directly to room temperature to form martensite. The patterns from martensite showed the expected broadening and shifting of diffraction lines relative to those of pearlite. In addition, the 'splitting' of lines diffracted at high Bragg angles indicated the tetragonality of the martensite structure. Patterns from the bainite structures were also diffuse, but less so, and no shifting or splitting of lines was apparent. The shapes of the [220] lines obtained from the three structures are shown in Fig. 6.

The broadening shown by the bainite and martensite lines is characteristic of diffraction from powders or crystallites of a 'particle-size' of less than about 1 000 Å, and from strained or distorted structures. For martensite structures, the analysis made by Wheeler and Jaswon[8] indicates that broadening is due to balanced internal stresses rather than to a small particle-size. A detailed analysis of this type

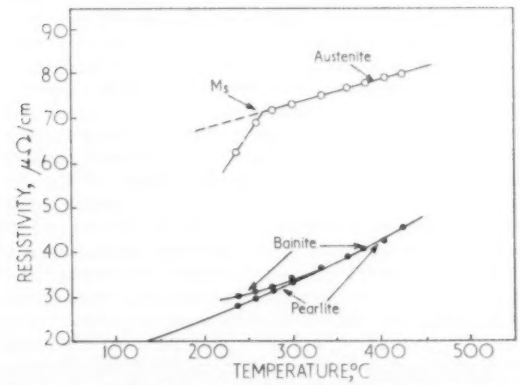


**Fig. 5—Changes in resistivity during isothermal reaction of iron–carbon alloy (0·77%C)**

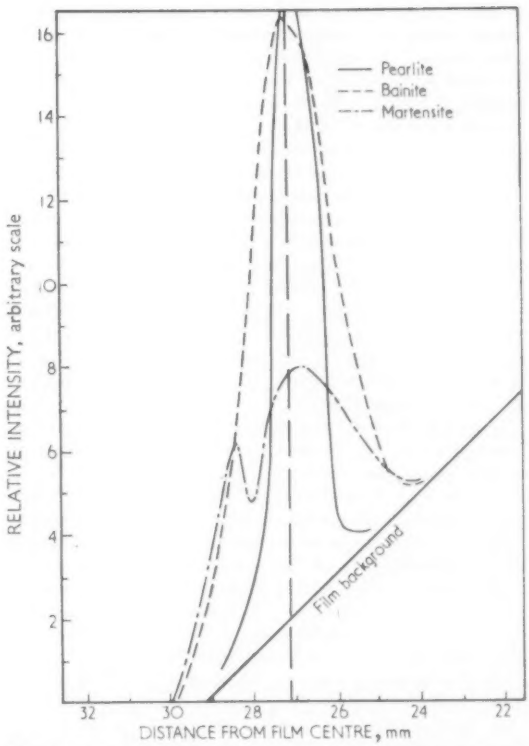


**Fig. 6—Appearance of [220] line on diffraction patterns for plain carbon steel (0·65%)**

has not been made for bainite. However, since examination of published electron-micrographs of bainite suggests that the size of ferrite regions is always greater than 1 000 Å, even in bainite formed at low temperatures, it is unlikely that the broadening observed in the bainite diffraction patterns is associated with a particle-size effect. Instead, it is concluded that the broadening is indicative of the existence of lattice strain within the bainite structure.

## DISCUSSION

### General features of the transformation

The transformation diagrams for the four materials exhibit the decrease in the rate of bainite formation

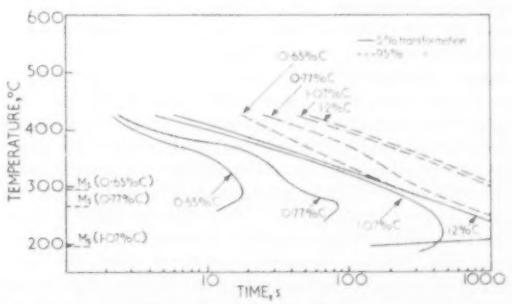


**Fig. 7—Comparison of times of 'start' and 'finish' of isothermal reaction**

at low temperatures which has been established by earlier studies of the reaction in plain carbon and alloy steels. In addition, at a given temperature, a decrease in rate with increasing carbon content is apparent. This reduction is shown clearly in Fig. 7 in which reaction 'start' and 'finish' times at the various temperatures have been plotted for all four materials. It has previously been held that an increase in carbon content decreases the rate of bainite formation but, as an examination of published transformation diagrams of commercial steels has shown, the actual effect is obscured by differences in austenitizing conditions and in the amounts of other elements present, in particular manganese. Only a general trend is shown by published diagrams; a high-carbon steel which is low in manganese may transform to bainite more quickly than one containing less carbon but more manganese.

It had been anticipated that the plain carbon steel examined in the present study would react more slowly than the eutectoid alloy. However, as is shown in Fig. 7 the reaction start for the steel falls into correct order with respect to carbon content. Since care was taken to maintain constant austenitizing conditions for all the materials, the probable explanation of this behaviour lies in the low manganese content of the steel. Whereas medium carbon steels contain normally some 0·5–0·8 wt-%Mn, the steel used here contained only 0·35%. Such a low level of manganese appears to have little effect on the kinetics of the bainite reaction.

At the lowest temperatures studied, the beginning of reaction shows in all cases a distinct 'swing-back' to shorter times. For the steel and the eutectoid alloy, the 'swing-back' occurs only at reaction temperatures below that at which martensite begins to form whereas for the hyper-eutectoid alloys, it begins above $M_s$. (The indicated $M_s$ temperatures were determined from the resistivity/temperature curves, as is shown below.) More rapid onset of the bainite reaction at temperatures below $M_s$ is a well established phenomenon and is associated with the accelerating effect of the presence of martensite plates on the nucleation of bainite in adjacent regions of untransformed austenite. Acceleration of austenite decomposition at temperatures just above $M_s$ is less well established, but has been observed in a plain carbon steel (1·35%C) by Howard[18] and in a number of alloy steels by Schaaber.[19] On the basis of kinetics and microstructure, Howard associated the acceleration with the formation, before that of bainite, of a new acicular constituent which might be free carbide. Schaaber, using dilatometric and inductive techniques, distinguished two-stage isothermal reactions in manganese, chromium, and silicon steels. The reactions generally overlapped in time, but in the case of a silicon steel, the initial reaction ceased before the second (bainite) reaction began. Schaaber concluded that the initial reaction might be the isothermal formation of martensite. The kinetics of the accelerated transformations observed here for the hyper-eutectoid iron–carbon alloys suggest that the later stages of transformation represent bainite formation, but no separation of a distinct first-stage reaction is apparent. Schaaber did not attempt, as did Howard, to identify microscopically the products of the initial reaction, and without such evidence the postulate of an isothermal martensite reaction is unjustified. Microscopic examination of specimens quenched to room temperature from the early stages of reaction was not feasible with the present transformation apparatus. Thus from the experimental observations it can be inferred that only at temperatures approaching $M_s$ the formation of bainite in the high-carbon alloys is preceded and accelerated by the formation of another constituent.

Measurements of the changes in electrical resistivity during continuous-cooling experiments have been used previously to determine $M_s$ temperatures in steels and other metals; the $M_s$ point is indicated by a discontinuity in the resistivity/temperature curves. In the present investigation, measurements were made of the resistivity of austenite immediately after quenching to the temperature of isothermal reaction. For temperatures below $M_s$, some martensite would be expected to form during cooling and the resistivity measured after quenching should be less, in proportion to the amount of martensite formed. Such behaviour was exhibited by the austenite resistivity curves for the materials examined, as is illustrated by Figs. 4 and 5. The point of intersection of the two discontinuous portions of each curve indicates the $M_s$ temperature. The values of $M_s$ obtained in this way confirm published values obtained by metallographic methods. Since $M_s$ temperature is markedly sensitive to variation in composition, this agreement provides further evidence of the accuracy of the resistivity method of determining the composition of pure iron–carbon alloys.

The resistivities of pearlite and bainite differ, in the case of the plain carbon steel, over the whole temperature range examined. The difference is fairly constant down to temperatures in the region of 450–350° C, below which it increases. Corresponding curves for pure alloys diverge only at temperatures below this range. In the case of alloy steels, a difference between the resistivity of pearlite and bainite structures would be expected from the work of Hultgren,[5] which demonstrated that in bainite (in contrast with pearlite) alloying elements do not partition between ferrite and carbide. It is not unreasonable to assume that this effect will also take place in plain carbon steels, even though the amounts of alloying elements present are small. The different distribution of alloying elements in the two structures would effect the electrical resistivities of the ferrite–carbide aggregates in the manner which has been found to occur for the plain carbon steel at temperatures above 300–350° C. However, this explanation does not suffice for the increasing divergence of the resistivities which are observed at lower temperatures for both the steel and the iron–carbon alloys.

At such temperatures, the internal strain existing within the bainite structure, which is evidenced by the broadening of X-ray diffraction patterns, would be accompanied by some increase in electrical resistivity above that of pearlite. However, the observed increase is considerable and it is doubtful whether strain alone could be the cause. The excessive dilation of specimens reacted to bainite could be

measured only approximately, but that the effect does occur in plain carbon steels has been confirmed recently[20] during a dilatometric study of bainite formation in a steel containing 0·97%C.

Although no change in the lattice parameter of ferrite in low-temperature bainite was discovered from X-ray measurements, the accuracy of such measurements is such that the effect of a carbon content of up to 0·1 wt-% might not be detectable. Dijkstra and Koster [21] have shown that precipitation of 0·01%C from solution in alpha iron leads to a decrease in electrical resistivity of 2·5%. If the higher resistivity of bainite over that of pearlite in the same alloy is attributed entirely to supersaturation of the ferrite in bainite with carbon, the degree of supersaturation at a given temperature can be calculated. In Table I the differences between the carbon contents of ferrite in bainite and in pearlite are presented for the three alloys. To avoid error due to the contribution from martensite to resistivity, differences were calculated only for temperatures above the $M_s$ temperature.

The amount of carbon retained at a given temperature is seen to be essentially the same in the three alloys, indicating that the degree of supersaturation depends only on the reaction temperature. Even at the lowest temperature for which calculation was possible, supersaturation does not exceed some 0·05 wt-%C.

Estimation of the change in length, assuming the ferrite to have a body-centred cubic structure, which should accompany the carbon retention indicated, gives values of the order necessary to cause the degree of 'bowing' of the wire specimens which was observed experimentally. Retention of carbon in solution would also be expected to give rise to the lattice strain evidenced by the broadening of diffraction patterns from bainite.

The identification of the carbide or carbides present in the bainite was impossible because of the weakness of the carbide lines on the X-ray diffraction patterns. However, the presence of $\epsilon$ carbide as well as cementite has been detected in bainite formed in a eutectoid plain carbon steel below 290° C.[4] If the bainite nucleus were a region of supersaturated tetragonal ferrite, of which the c/a ratio decreased as carbide was precipitated within the growing region, then at temperatures where $\epsilon$ carbide is formed the equilibrium carbon content of the ferrite should correspond to that obtained after completion of the first stage of the tempering of martensite, i.e. to some 0·3%C. The possible presence of such regions of tetragonal ferrite suggests an alternative explanation of the high electrical resistivities of bainite structures formed below 300–350° C. However, using an X-ray technique which is more sensitive than that used here, Werner *et al.*[22] have demonstrated recently that in bainite formed at 210° C in a pure iron–carbon alloy (1·43%C) ferrite has the same lattice structure and parameters as pure iron. In contrast, martensite tempered for as much as 10 h at 210° C retains some tetragonality. Although the experimental uncertainty of Werner's measurements was such that the effect on the lattice parameters of the presence of up to 0·1%C in solid solution would not be detected,

TABLE I

**Carbon contents in ferrite, bainite, and pearlite, for three alloys**

| Carbon content of alloy, wt-% | Temperature, °C | Difference in carbon content between ferrite in bainite and ferrite in pearlite, wt-% |
|---|---|---|
| 0·77 | 265 | 0·016 |
| 1·07 | 265 | 0·018 |
| | 200 | 0·053 |
| 1·2 | 265 | 0·016 |
| | 180 | 0·053 |

the results do indicate that the high electrical resistivity of lower bainite is unlikely to be due to the presence of regions of tetragonal ferrite containing $\epsilon$ carbide. Instead, any $\epsilon$ carbide formed must be in equilibrium with cubic ferrite of low carbon content, as occurs on the quench-ageing of iron–carbon alloys.

### Analysis of isothermal reaction data

Two types of rate equation have been applied to the experimental reaction data obtained for the plain carbon steel and the three iron–carbon alloys. With the inclusion of impingement factors, the logarithmic forms are:

$$\log . \log\left(\frac{1}{1-y}\right) = n \log t - n \log k - 2{\cdot}3 \text{ (Zener)}^{23} \quad (1)$$

$$\log \frac{y}{1-y} = n \log t - \frac{k}{2{\cdot}3} \text{ (Austen and Rickett)}^{9} \quad \ldots(2)$$

Where $y$ is the fraction of austenite transformed at time $t$, $k$ is a temperature dependent rate-constant, and $n$ is a constant.

Experimental data which conforms to these equations will lie on a straight line when plotted in the form of $\log . \log\left(\frac{1}{1-y}\right) / \log t$ for equation (1), or $\log\left(\frac{y}{1-y}\right) / \log t$ for equation (2).

Application of the Austin-Rickett equation to the reaction data obtained in the present study has shown that, in general, the equation fits the results for any particular reaction to a higher proportion of transformation than does the Zener equation. However, at the later stages of transformation the equation was found to overcompensate for the fall in reaction rate, consequently its use was discontinued and in the analysis to be described only the Zener equation has been used.

The isothermal reaction data for the 1·07%C alloy are shown in Fig. 8 plotted in accordance with equation (1). The curves are representative of those obtained from all four materials. With the exception of reactions in the vicinity of the $M_s$ temperature, the experimental data conforms to equation (1) up to some 50–70% of the total transformation. Beyond this, the curves fall away from the straight lines drawn through the data points for the previous part of the reaction. The accelerated curves obtained near to $M_s$ conform in their middle stages to a linear plot of the type found for the higher temperature reactions. As the transformation temperature is

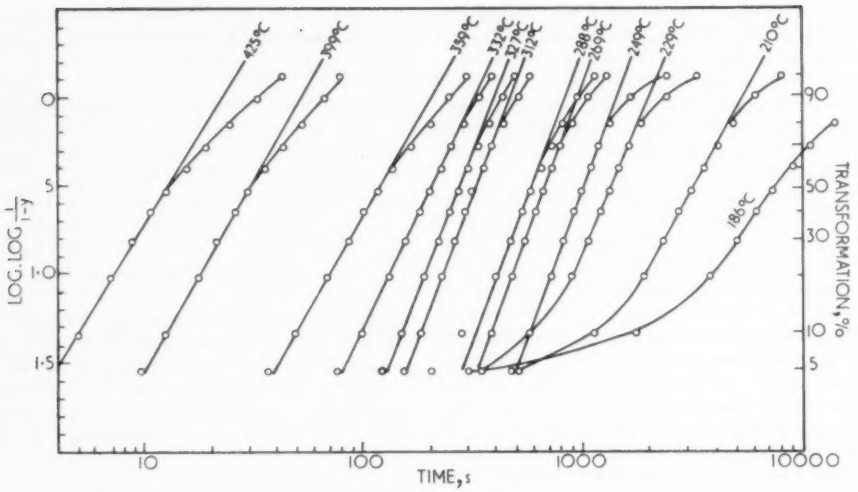


**Fig. 8—Derived reaction-rate curves for iron–carbon alloy (1·07%C)**

lowered into the martensite region, the linearity of the middle stage becomes increasingly affected.

Values of the index $n$ were obtained from the slopes of the linear portions of curves which did not exhibit an accelerated stage, and these are shown in Fig. 9 as a function of the temperature of isothermal reaction. The value of $n$ changes in a distinctive manner with temperature. The indices for reaction temperatures below 300° C all lie within the range 3–4, whereas those for temperatures above about 350° C lie within the range 1·8–2·6. In the temperature range between, the values change from those characteristic of the upper temperature to those of the lower. That the change is real and not a spurious effect associated with the experimental technique is evidenced by the fact that defects in technique which might become more apparent with increasing temperature, such as imperfect quenching or failure to maintain the temperature of the specimen constant during the reaction, would be expected to increase the rate of the transformation and, therefore, the magnitude of the index $n$.

Optical- and electron-microscope studies of bainite structures have shown that the precipitates are not uniformly of the simple shapes considered in theoretical analysis of reaction rate. In addition, in the type of plot shown in Fig. 8 small differences in slope are reflected as substantial differences in the numerical value of the gradient. For these reasons, the use of the nearest 'ideal' indices within the two main ranges is considered to be justifiable. Thus the possible values of the index are 2·0 or 2·5 for the upper temperature range and 3·5 for the lower range.

No significant trend in the magnitude of $n$ with variation in composition is apparent in the upper temperature range. Although the movement to higher values as the temperature is lowered below 350° C does appear to be less marked for the steel and the eutectoid alloy than for the two higher carbon alloys, the proximity of the $M_s$ temperature of the two former materials makes it difficult to discern a definite effect of composition.

The temperature dependence of the reaction constant $k$ in equation (1) can be expressed in the form derived empirically by Arrhenius as:

$$\frac{1}{k} = C \cdot \exp\left(-\frac{Q}{RT}\right) \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (3)$$

where $Q$ and $C$ are constant, $T$ is the absolute temperature, and $R$ the gas constant. Experimental data which satisfies this equation will lie on a straight line when plotted in the form of log $(1/k)$ $v.$ the reciprocal of the absolute temperature. The slope of the line is a measure of $Q$, which has the dimensions of energy and, if $R$ is taken to be 1·98 cal,

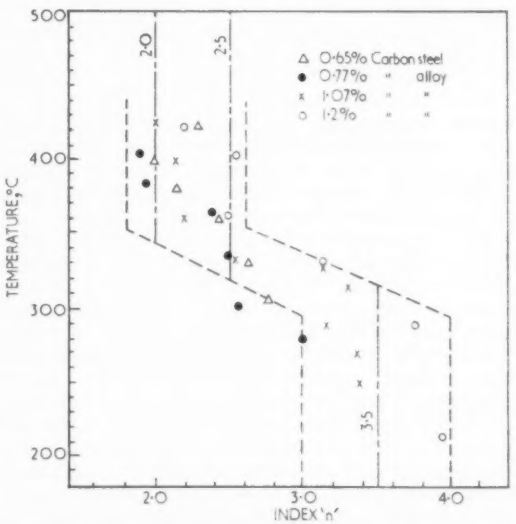


**Fig. 9—Effect of transformation temperature on the index $n$**

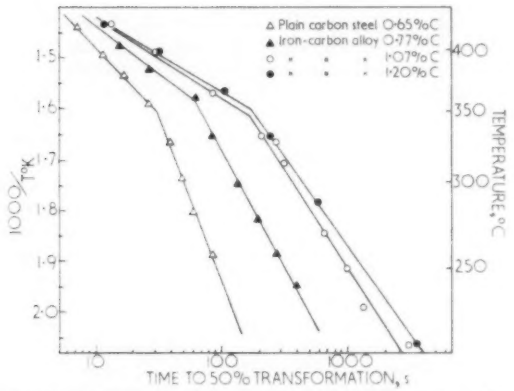


**Fig. 10—Arrhenius plot of reaction data for the plain carbon steel and iron–carbon alloys**

is measured in cal/g-mol. The quantity $Q$ is generally regarded as an 'activation energy', that is, an energy level which individual atoms must exceed if they are to take part in the reaction process. Thus, the rate of the overall process is proportional to the number of atoms whose energy exceeds the value of $Q$. The function $\exp(-Q/RT)$ varies rapidly with $Q$, and since characteristic activation energies are associated with different processes, any change in the nature of the reaction with temperature is reflected by a change in the slope of the graph of $\log(1/k)$ *v.* $1/T$. In the case under consideration, it is more convenient to plot $\log t$ instead of $\log(1/k)$, where $t$ is the time for a definite fraction of the reaction to be accomplished. The equation which justifies this substitution can be derived from insertion of the Arrhenius relationship for the constant $k$ into equation (1).

The reaction data for the steel and the pure alloys are shown in Fig. 10 plotted in the form of $\log t_{50}$ *v.* 1/T, where $t_{50}$ is the time to reach 50% transformation. For each material, the data fall on two straight lines which intersect in the region of 350° C. Thus, the indications from the changes in the electrical resistivity and dilational characteristics and in the shape index $n$, that there are marked differences between the bainite reactions above and below about 350° C are strikingly confirmed. The values of $Q$ obtained from measurements of the slopes of the lower and upper portions of the lines in Fig. 10 are given in Table II. These experimental 'activitation energies' are expressed as a function of carbon content in Fig. 11.

**TABLE II**
**Values of 'Q'**

| Material | wt-%C | Q, cal/mol (± 1 000 cal) | |
|---|---|---|---|
| | | Above 350° C | Below 350° C |
| Plain carbon steel | 0·65 | 18 500 | 7 500 |
| Iron-carbon alloy | 0·77 | 22 000 | 9 500 |
| ,, ,, ,, | 1·07 | 29 000 | 11 500 |
| ,, ,, ,, | 1·20 | 32 000 | 13 000 |

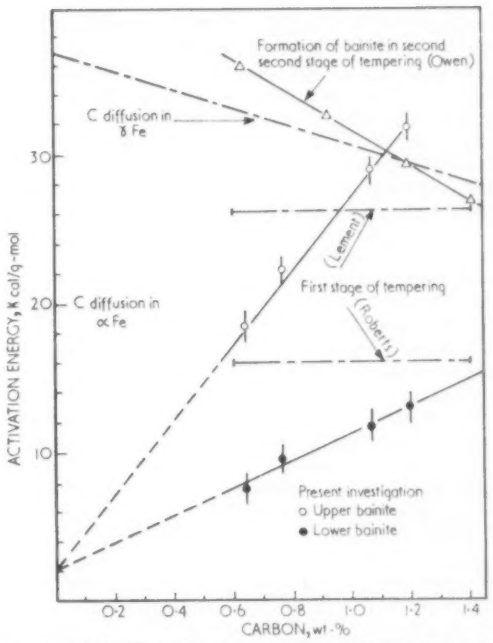


**Fig. 11—Relations between carbon content and activation energies for different processes**

The kinetics of bainite formation thus support the conclusion which can be drawn from electron-microscope observations of bainite structures—that a distinct change in the characteristics of isothermally formed bainite occurs over the range of reaction temperatures from about 350–300° C. It is apparent that the different microstructures of upper and lower bainite are associated with reaction processes which differ considerably in their kinetics.

The fact that the upper limit of the lower bainite reaction lies in the above temperature range for all the carbon contents examined, suggests that for medium alloy steels the $B_f$ temperature, which falls in most cases within the same temperature range, may be associated with the change to a lower-bainite type of reaction which enables the transformation to go to completion.

### General discussion

Comparison of 'activation energies' obtained from analysis of reaction curves, with those for the diffusion of the various atoms present in a system is commonly used to determine the diffusion process which controls the reaction rate. In Fig. 11, the known activation energies for the diffusion processes which could be concerned in the bainite reaction are indicated, with the exception of that for the self-diffusion of iron, about 70 000 cal/mol, which is too large to be included.

It has been suggested that the diffusion of carbon in austenite is the most likely controlling process in the formation of bainite. The suggestion has received some support from determination of overall activation energies for the second stage of tempering, which is the formation of bainite from austenite retained by

quenching.[25] The energies for this reaction have also been included in Fig. 11 and can be seen to conform fairly well with those for carbon diffusion in austenite. The value obtained for the index $n$ from the tempering kinetics was 1·6, which led to the conclusion that bainite grew as plates restricted by edgewise impingement and in a manner dependent on both steady-state nucleation and growth. These results differ markedly from those obtained here for the direct formation of bainite. However, it is not considered that the results are incompatible since the conditions under which bainite forms are very different in the two cases. The effect of internal strain as the result of quenching, and the presence of martensite would be expected to influence considerably the transformation of retained austenite to bainite.

The fact that the overall activation energies found in the present investigation decrease with decreasing carbon content, suggests that the controlling process is not that of carbon diffusion through austenite. Since the decrease extends to below the values for carbon diffusion in ferrite and in martensite, neither of these processes can be rate-determining. It is apparent therefore, that the formation of bainite is not controlled by a known diffusion process. The only determination of rates of growth of bainite plates has been made by Tsuya[7] for a forging die steel (0·76%C–2·16%Ni–1·14%Cr–0·24%Mo). The rates were measured from observation in a hot-stage microscope of surface relief markings formed on specimens during isothermal transformation in the range 275–425° C. A value of 19 000 cal/mol was obtained for the activation energy of the process and from this Tsuya deduced that carbon diffusion in ferrite controls the growth of bainite. However, the considerable scatter of the data indicates that the probable error in the energy value is $\pm$ 2 500 cal/mol. In view of this, and of the influence of carbon content on the overall reaction energy which has been established for non-carbon alloys, it is doubtful whether Tsuya's conclusion is valid. Similar growth measurements on iron–carbon alloys are required to determine the part played by growth in the bainite reaction.

The linear manner in which the experimental activation energies increase with increasing carbon content from a common extrapolated value at zero carbon content to a value of the order for that for carbon diffusion in austenite in the case of upper bainite, and of that for carbon diffusion in martensite in the case of lower bainite, has obvious significance. The common extrapolated value, within the limits of the accuracy of measurement, is close to that for the formation of martensite plates which occurs by a nucleation and shear process. The deduction can be made that both the formation of regions of supersaturated ferrite from the austenite matrix, and the diffusion of carbon through austenite for upper bainite, (or through supersaturated ferrite for lower bainite,) are important in controlling the overall reaction. At low carbon contents the reaction is dominated by the transformation to ferrite, and as the carbon content is increased the removal of carbon from the ferrite and from the austenite in advance of the region becomes increasingly important in allowing transformation to proceed. The removal can occur by diffusion into the austenite away from the region, or to carbide particles precipitated within the ferrite. Although these deductions account qualitatively for the energy changes observed experimentally, it has been found impossible to develop a quantitative model of the reaction on this basis. In the present state of reaction-rate theory, it is difficult to deal with reactions which are not controlled by a single process.

The slower start of transformation at all temperatures as the carbon content of the alloy is increased suggests that the rate is concentration-dependent. The manner of the time shift with increasing carbon content of the start of the reaction confirms the view that the nucleus for bainite formation is ferrite. Increasing the carbon content will reduce the free-energy difference between austenite and ferrite existing at the same temperature and thus diminish the driving force for the nucleation of ferrite, making its precipitation less probable. In contrast, the driving force for, and the probability of, cementite precipitation from austenite increases with increasing carbon content.

It is considered to be unjustifiable to attempt a detailed account of bainite formation without a knowledge of the nucleation or growth characteristics of the reaction. However, it is possible by the addition of what has been ascertained during the present study to the facts established previously, to suggest the general nature of the transformation of austenite to bainite in the materials examined. The precipitation process which is observed at any temperature at which austenite is in a metastable condition, will be the one that occurs at the maximum rate at that temperature. Throughout the temperature range examined, the formation of bainite is preceded by the precipitation of ferrite which is coherent with the austenite matrix. In the upper bainite range the reaction proceeds by the precipitation of carbide within the ferrite regions which, as the result of carbon diffusing from the austenite to the particles of carbide, grow into the austenite matrix. At temperatures in the lower bainite range it becomes possible to increase the rate of transformation by altering the shape and composition of the carbide and the composition of the ferrite. An alteration in the carbide composition is suggested since a condition of metastable equilibrium with a higher carbide than cementite ($Fe_3C$) would account for the increased carbon content of the ferrite observed.

There is some evidence from studies of the second and third stages of tempering in steels, that carbon-rich ferrite may exist in a metastable equilibrium with free carbide. Lement, Averbach, and Cohen[26] concluded that coherency between cementite and the ferrite matrix is unlikely to be the sole cause of the enhanced solubility of carbon and that the existence of a carbide other than cementite could not be discounted.

In the lower bainite range, growth of bainite regions takes place by diffusion of carbon through the ferrite to the carbide particles contained within it. It is possible that a change in nucleation conditions is solely responsible for the changes in transformation rate observed on passing from the upper to the lower range. It is considered, however, to be more likely

that in the lower range, growth proceeds more rapidly than carbon can diffuse through austenite, and that the growing bainite region sweeps up carbon as it advances. Thus the rate of growth of the region depends on the transport of carbon through supersaturated ferrite.

## CONCLUSIONS

(1) A marked difference exists between the kinetics characteristic of the isothermal transformation of austenite to upper bainite and those of the transformation to lower bainite.

(2) Based on the changes observed in kinetics and electrical resistivity, the transfer from the upper to the lower bainite reaction takes place in the temperature range 300–350° C.

(3) Accompanying the formation of lower bainite, there is an increase in the carbon content of the ferrite in the bainite aggregate. The increase is possibly associated with a change in the composition of the carbide.

(4) The formation of bainite over the temperature range 200–450° C is not a simple, growth-controlled reaction.

(5) The transformation of austenite to bainite is considered to be initiated by the precipitation of ferrite at a rate which is both time and composition dependent.

### Acknowledgments

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### References

1. V. G. Paranjpe and D. D. Kaushal: *Trans Indian Inst. Met.*, 1951, vol. 5, pp. 147–172.
2. R. F. Hehemann and A. R. Troiano: *Met. Prog.* 1956, vol. 70, No. 2, pp. 97–104.
3. Progress Reports of the Committee on Metallography: *Proc. ASTM*, 1950, vol. 50, pp. 444–454; 1952, vol. 52, pp. 543–572; 1953, vol. 53, pp. 505–530.
4. A. E. Austin and C. M. Schwartz: *Ibid.*, 1952, vol. 52, pp. 592–595.
5. A. Hultgren: *Trans ASM*, 1947, vol. 39, pp. 915–1 005.
6. T. Ko and S. A. Cottrell: *JISI*, 1952, vol. 172, pp. 307–313.
7. K. Tsuya: *J. Mech. Lab.* 1956, vol. 2, No. 1, pp. 20–27.
8. N. P. Allen, L. B. Pfeil, and W. T. Griffiths: *Iron Steel Inst. Spec. Rep.* No. 24, pp. 369–390: 1939.
9. J. B. Austin and R. L. Rickett: *Trans AIME*, 1939, vol. 135, pp. 396–415.
10. O. Krisement and F. Wever: *Inst. Met. Symposium*, 'The Mechanism of Phase Transformations in Metals', 1955, pp. 253–263.
11. S. V. Radcliffe and E. C. Rollason: *JISI*, 1958, vol. 189, pp. 45–48.
12. A. W. McReynolds: *J. Appl. Phys.*, 1946, vol. 17, pp. 823–833.
13. F. Wever and W. Jellinghaus: *Mitt. K. W. Inst.*, 1933, vol. 15, pp. 167–177.
14. W. H. Colner and O. Zmeskal: *Trans ASM*, 1952, vol. 44, pp. 1 158–1 168.
15. C. West: *ASM* Seminar 'Thermodynamics in Physical Metallurgy', 1950, pp. 178–200.
16. T. F. Russell and C. Mavrocordatos: *JISI*, 1949, vol. 162, pp. 33–43.
17. J. A. Wheeler and M. A. Jaswon: *Ibid.*, 1947, vol. 157, pp. 161–166.
18. R. T. Howard and M. Cohen: *Trans AIME*, 1948, vol. 176, pp. 384–400.
19. O. Schaaber: *J. Met.*, 1955, vol. 7, No. 4, pp. 559–560.
20. S. W. Graham: Private communication from Manchester University.
21. L. J. Dijkstra: *Philips Res. Rep.* 1947, vol. 2, No. 5, pp. 357–381.
    W. Köster: *Arch. Eisenhütt*, 1928, vol. 2, pp. 503–522.
22. F. E. Werner, B. L. Averbach and M. Cohen: *Trans AIME*, 1956, vol. 206, p. 1484.
23. C. Zener: *J. Appl. Phys.*, 1949, vol. 20, pp. 950–953.
24. W. A. Johnson and R. F. Mehl: *Trans AIME*, 1939, vol. 135, pp. 416–442.
25. W. S. Owen: *Trans ASM*, 1954, vol. 46, pp. 812–828.
26. B. S. Lement, B. L. Averbach and M. Cohen: *Ibid.*, 1954, vol. 46, pp. 851–877.
27. J. C. Fisher: *ASM* Seminar 'Thermodynamics in Physical Metallurgy', 1950, pp. 201–241.
28. C. Zener: *Trans AIME*, 1946, vol. 167, p. 550.
29. R. F. Hehemann and A. R. Troiano: *J. Met.*, 1954, vol. 6, Nov., pp. 1 272–1 280.

# *Powder Metallurgy Joint Group*

## Symposium on ceramic–metal materials

As announced in the December issue of the *Journal* (p. 393), the Powder Metallurgy Joint Group of The Iron and Steel Institute and The Institute of Metals held a symposium on "The Powder Metallurgy of Ceramic–Metal Materials" on 16th and 17th December. Abstracts of the six main papers in the symposium are given here. They will be published in full in the Group's new journal, *Powder Metallurgy*.

"*Recent developments in the field of silicides and borides of the high melting point transition metals*" by R. Kieffer and F. Benesovsky (Metallwerk Plansee A.G., Reutte (Tirol), Austria).

The binary systems of silicides of the high melting point transition metals are now well understood, except for the hafnium–silicon system. Research since 1954 is reviewed, with particular reference to the compound $Me_5Si_3$ and its position in the silicide systems. Reference is also made to the pseudo-binary and pseudo-ternary silicide systems.

The structures of many of the intermetallic phases in the binary boride systems have now been determined, but complete equilibrium diagrams still remain to be established in some cases. New tentative diagrams are given for the systems vanadium–boron, niobium–boron, and tantalum–boron and structures are suggested for the borides $V_3B_2$, $Nb_3B_2$, and $Ta_3B_2$ with the T2 structure (isostructural with $U_3Si_2$).

Ternary alloys of the systems Me–Si–B are of great interest, not only structurally but also for practical reasons. The complete systems Me–Si–B of Group VI (Mo–Si–B and W–Si–B) have accordingly been studied by X-ray, thermal-analysis, and micrographic methods. The system Cr–Si–B has been determined and attention is directed to the possible commercial applications of certain alloys containing additions of metals of the iron group, in particular nickel, for sprayed coatings resistant to liquid aluminium.

The question of cementing silicides and borides with metals and alloys is discussed theoretically. The character of the silicide or boride system in question and the behaviour of the intermediate phase in relation to the bonding material are of decisive importance for the selection of the latter. Only very limited data are to be found in the literature on the behaviour of silicides and borides in relation to metals and alloys. Alloys based on $TiB_2$, $ZrB_2$, $MoSi_2$, and $WSi_2$, impregnated with numerous metals and alloys, have been prepared. Their structures have been studied and the technical suitability of various combinations is discussed on the basis of their technological properties.

"*Bonding in carbides, silicides, and borides*" by D. A. Robins (Research Laboratories of the General Electric Co. Ltd., Wembley).

The hypothesis that many of the properties of the transition metals, their alloys, and their compounds can be understood in terms of a relationship between the co-ordination of the metal atoms and the number of bonding electrons per atom, is applied to the carbides, silicides, and borides of the transition metals of Groups IVA, VA, and VIA. The crystal structure and stability of the carbides can be satisfactorily explained on the basis of electron concentration, but a complete understanding of the bonding in the silicides and borides is more difficult. In the silicides the metal–silicon bonding, which is metallic in character, becomes of increasing importance as the silicon content is increased, and in the disilicides no direct metal–metal bonding remains. The ability of boron to form strong boron–boron bonds is important in the borides, and the metal–metal co-ordination would appear to be of only secondary importance. The practical significance of the principles put forward is discussed.

"*The high-temperature properties of ceramics and cermets*" by E. Glenny and T. A. Taylor (National Gas Turbine Establishment, Farnborough).

It has been recognized for some time that, with progressive increase in the working temperature of gas turbines, metallic alloys may no longer be adequate for rotor or stator blading. The use of more refractory but more brittle materials, i.e. ceramics and ceramic–metal mixtures (cermets), has been suggested. The paper is concerned with the evaluation of the major properties involved, namely creep strength, fatigue strength, resistance to thermal fatigue (i.e. to repeated thermal shocks), oxidation-resistance, and impact-resistance. The materials evaluated include oxides, oxide–metal cermets, carbides, carbide–metal cermets, molybdenum disilicide, and silicon nitride. The equipment for determining the effects of alternating and steady mechanical stresses up to 1200° C is described. The relative merits of the test materials are discussed. It is concluded that the resistance to thermal fatigue and to impact of the ceramics and cermets is inferior to that of metallic alloys in current use.

"*The oxidation of zirconium carbide in high-temperature combustion gases*" by W. Watt (Royal Aircraft Establishment, Farnborough).

The reactions of sintered zirconium and titanium carbides have been studied in simulated rocket combustion gases at surface temperatures of about 2 250° C. Under these conditions zirconium carbide is oxidized according to a parabolic law owing to the formation of an adherent oxide film. No oxide film is formed on titanium carbide under the same conditions, and it decreases in weight and dimensions at a relatively rapid rate.

"*The fabrication and properties of chromium–alumina and molybdenum–chromium–alumina cermets*" by J. B. Huffadine, L. Longland, and N. C. Moore (The Plessey Co. Ltd., Towcester).

Following a brief review of the literature, various methods for preparing chromium–alumina and molybdenum–chromium–alumina cermets are described. The properties of such materials are discussed and an indication is given of their possible applications.

"*The fabrication and properties of uranium oxide–iron cermets*" by W. J. Wright, R. B. Gibbon, and J. Williams (Atomic Energy Research Establishment, Harwell).

Cermets of iron–uranium dioxide, containing up to 30 vol-% $UO_2$, have been rolled to rod. Although the cores rolled centrally along the length, the core was distorted in section owing to stiffening of the core relative to the sheath as the temperature dropped during rolling. The $UO_2$ particles were fragmented during rolling and formed as 'stringers' along the direction of rolling, seriously weakening the structure transversely.

By extrusion at 1 200° C, cermets containing up to 50 vol-% $UO_2$ were satisfactorily fabricated. The extrusion pressure increased sharply above 35 vol-% of ceramic and when large reduction in area were taken. The extrusion characteristics do not conform to isotropic plastic extrusion nor to pure viscosity effect; the influence of the dispersed particles on metal flow is explained by a strong interference effect between particles and the metal during flow. The properties of the extruded rods were good; the core density was 90–95% of theoretical, decreasing slightly with increasing ceramic content; there were no obvious directional properties in the core, nor was the ceramic fragmented as in rolling. The characteristic 'tail-end defect' of the extrusions was partly corrected by using shaped sealing plugs.

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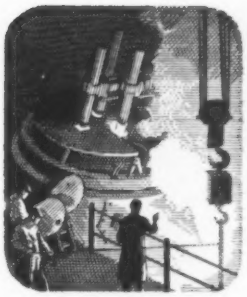

# Vacuum pouring of ingots for heavy forgings

*By J. H. Stoll*

SYNOPSIS

The vacuum pouring process is indicated as a major advance in the industry for the production of large heavy forgings. It involves melting of steel in conventional furnaces and subsequently pouring into an ingot mould in a vacuum tank. Reduction of hydrogen content and protection of the molten steel from oxidation during pouring result in important improvements in the quality of the product. The experimental use of a 7-ton capacity unit, coincident with the construction of a 250-ton production unit at the Bethlehem Steel Company enabled the development of practices and techniques for immediate application to the large unit on initial production. Ingots can be produced as large as 120 in. dia. and 500 000 lb weight. The advantages of the process are obtained with considerable expense for equipment and operation. However, the improvement in quality of product more than justifies the added cost for critical applications of forgings. 1644

IN COMMON with most other products of the industrial world, the modern trend in the heavy forging field has been toward a bigger and better product. The manufacture of high-quality large forgings has for many years presented a challenge to steel producers. The increasing intensity of high quality requirements, higher mechanical properties, and additional property specifications have served to increase this challenge. There have been created some most difficult problems in obtaining sound forgings of large dimensions for highly stressed applications such as those of the electric power, atomic energy, and other special industry requirements.

Vacuum pouring of steel shows great promise towards the solution of some of these problems. It is the purpose of this paper to describe this process briefly as carried out at Bethlehem and to discuss experience there, reviewing the potential benefits of the process, as indicated by the limited production up to the present time.

Heavy forgings, in our terms, are produced under a hydraulic press with open die forging. Depending upon their size and weight, they are forged from rich alloy composition ingots up to 120 in. dia. having a maximum weight of 500 000 lb, and certain carbon-steel or low-alloy ingots as large as 134 in. dia, having a maximum weight of 700 000 lb. Typical compositions used in the U.S.A. for these forgings are:

| | *Element, %* | | | | | | |
|---|---|---|---|---|---|---|---|
| *Type* | *C* | *Mn* | *Si* | *Ni* | *Cr* | *V* | *Mo* |
| C–V | 0·40 | 0·65 | 0·20 | | | 0·05 | |
| Ni–V–Mo | 0·25 | 0·65 | 0·20 | 2·85 | | 0·07 | 0·50 |
| Ni–Cr–V–Mo | 0·30 | 0·70 | 0·20 | 2·85 | 1·05 | 0·25 | 0·55 |
| Cr–V–Mo | 0·30 | 0·85 | 0·25 | | 1·05 | 0·25 | 1·25 |
| Ni–Cr–Mo | 0·20 | 0·65 | 0·25 | 0·70 | 0·35 | | 0·60 |

Problems involved in the successful production of these large ingots and resulting from subsequent processing and handling may well be expected to tax to the limit, not only the necessary manufacturing facilities, but also the manufacturing 'know-how' gained over years of experience.

The problems concerned with these large masses of steel are many and varied, but some of the most troublesome are associated with internal soundness.

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For the production of heavy forgings it is necessary to start with a large ingot which follows all the natural laws of solidification and, at best, will contain the normal segregation and porosity associated with the solidification of large masses. The ingot is forged not only to obtain the desired size and shape, but also to gain the advantage of the mechanical work which is beneficial in breaking up these segregates and in consolidating the normal ingot porosity The forging must then be given a long thermal conditioning cycle to refine the coarse structure resulting from the high temperature of forging and to prevent the occurrence of flakes. Subsequently, the forging is heat-treated to attain the desired mechanical properties.

After these various steps have been accomplished, modern inspection methods make it possible literally to peer into the forging and thoroughly examine it for internal soundness. Fortunately, a large number of the forgings produced come through with flying colours. However, all too often, in spite of the use of the best practices known to the forging industry, we are confronted with a large forging, representing considerable dollar value, that does not meet our rigid inspection standards nor those of our customers. When these forgings are subsequently subjected to metallurgical investigation, it is quite often found that flaking is responsible for the difficulty, even though all possible precautions, including the use of long and intricate thermal cycles, have been exercised throughout processing. With ever-increasing forging size and alloy content, the practice of preventing flaking by methods of hot handling and extended thermal cycles soon reaches the point of diminishing returns.

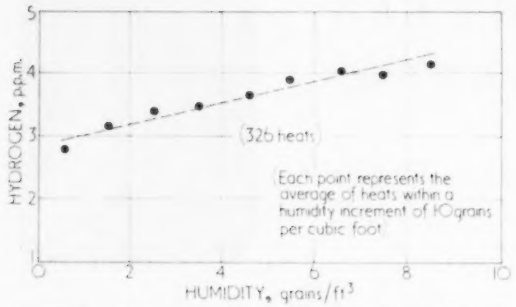

HYDROGEN, p.p.m.
HUMIDITY, grains/ft³
(326 heats)
Each point represents the average of heats within a humidity increment of 10 grains per cubic foot

**Fig. 1—Pony ladle hydrogen versus humidity**

## SOURCES OF HYDROGEN

Since hydrogen is generally accepted as a necessary factor in the formation of flakes, it becomes apparent that for successful production of large highly alloyed forgings the hydrogen content must be reduced. To accomplish this, the source of the hydrogen must first be recognized and a method then developed for either keeping it out of the steel or eliminating it before it can cause flaking.

Hydrogen may be introduced into the steel during normal melting operations through moisture from the raw materials, such as ferro-alloys, scrap, and slag-making ingredients. Ladle and hot-top refractories, as well as the ingot mould itself, if not thoroughly dried, can also contribute to the hydrogen content. All these factors are recognized and most modern melting shops are equipped to do the necessary drying of additions, ladles, hot-tops, and moulds. Perhaps not so well recognized is the fact that throughout the melting and pouring operation the steel is surrounded by a blanket of air carrying amounts of moisture, depending on the humidity at the time.

Figure 1, representing tests from 326 heats, shows that the hydrogen content of the steel is related to the moisture content of the air during melting and pouring. As the moisture in the air increases, hydrogen increases in the steel. It has been our experience that the highest-hydrogen heats can be anticipated in hot, humid weather.

## VACUUM PROCESSES

If air could be excluded during the melting and pouring operations a major source of hydrogen would be eliminated. This is possible, but only in limited quantities of steel, under present-day vacuum-melting processes. Vacuum-induction and consumable-electrode methods provide two commercial processes for melting steel under vacuum, and both are capable of producing material with low hydrogen content. Unfortunately, these methods are limited at the present time to low quantity production, about 2 500 lb for vacuum-induction melting and 12 000 lb for the consumable-electrode process. Obviously, neither of these methods is practical for adaptation to the production of a 120-in. dia. 500 000-lb ingot.

Figure 2 contrasts the size of ingots required for the larger utility forgings and an ingot representing the maximum weight producible by present vacuum-melting installations. The 120-in. ingot is one of the largest sizes made by vacuum pouring, while the 24-in. ingot is typical of the larger sizes obtainable by vacuum melting or consumable-electrode process.

As previously mentioned, if hydrogen cannot be kept down to the desired level during melting, a method must be found for subsequently reducing it, to prevent it becoming a factor in the formation of flakes. Subjecting the steel to a vacuum during pouring provides the best approach to this objective.

The pouring of metal *in vacuo*, or the use of low-pressure treatment of molten metal, has been suggested many times in the past as a means of reducing the gas content of metals. In fact, literature surveys disclose a method of pouring ingots in vacuum as early as 1883.* It was not, however, until the relatively recent development of vacuum pumps capable of producing vacua in volume at the micron level that serious consideration was given to the possibility of vacuum pouring of large tonnages.

## 7-TON DEVELOPMENT UNIT

Early in 1956, Bethlehem decided to pursue the possible advantages of vacuum pouring for large ingot production. Recognizing the multiplicity of problems involved in the successful operation of a large 250-ton unit, it was decided to design and install a development unit capable of taking the product of a 7-ton electric furnace. Completion and initial operation of this size of unit was feasible long before complete installation of the 250-ton unit could be accomplished.

The general arrangement of this unit is shown in Fig. 3. The development unit is capable of handling production-size ingots up to 40 in. dia. It proved of immeasurable value in working out the mechanics of the procedure and, at the same time, permitted evaluation of certain of the benefits of vacuum pouring. The design of this unit will not be discussed in detail, since it is identical with the 250-ton unit, to be described later.

The 7-ton development unit more than fulfilled its intended purpose of providing necessary information, especially from the standpoint of the mechanical features involved in operating a vacuum-pouring unit. By its judicious application to the process, much

* Robert H. Gordon: U.S. Patent 287273, 23rd Oct., 1883.

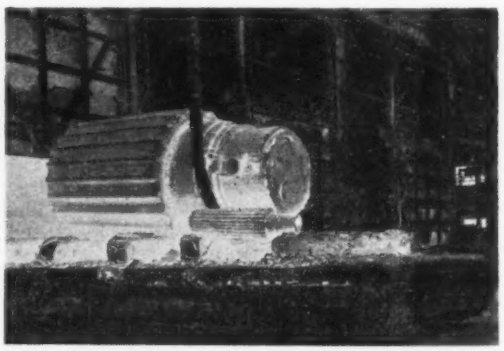


**Fig. 2—Contrast in size of vacuum-poured 120-in. ingot with 24-in. ingot, typical of larger-sized vacuum-melted ingot**

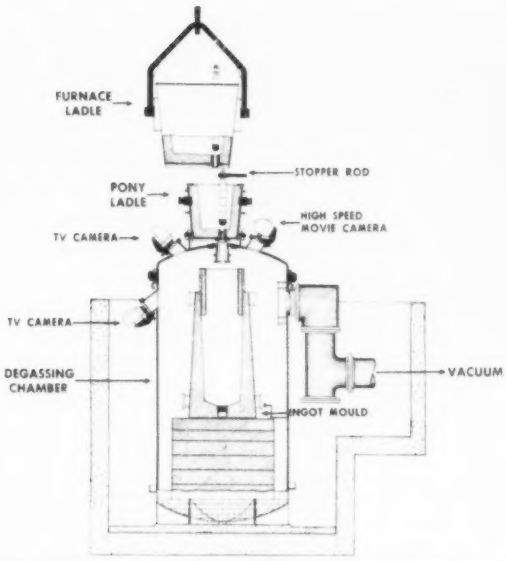



**Fig. 3—Schematic diagram of 7-ton vacuum-degassing unit**

experimentation was avoided when the larger unit became available, making it possible to put the large unit in operation on a production basis immediately. In this connection it is of interest to note that the first pour in the large unit was used successfully as an ingot for a production order.

In addition to providing the necessary mechanical 'know-how' for operation of the vacuum unit and the valuable training of personnel, the production and study of over 30 heats led to a number of preliminary conclusions regarding the vacuum process:

(1) An average of 60% of the hydrogen was removed, as indicated by a molten metal sample taken from the sinkhead of the ingot and compared with the pony-ladle analysis.

(2) The sensitivity to flaking was greatly decreased, as measured on the limited masses produced from a 40-in. ingot.

(3) The actual operating pressure must be low but may vary considerably in the low range. Hydrogen contents, as determined on molten metal from the sinkhead, were as low as 1 and 1½ p.p.m. obtained over a pouring range of pressures up to 1200 μ.

(4) Exceedingly valuable information was obtained about the contour of the spraying molten stream during vacuum pouring. It is indicated that the stream fans out to nearly 45° from the vertical during pouring.

(5) Information was obtained about the cone formed under the nozzle, as a result of the fanning of the stream. If the distance between the nozzle and the top of the mould is too great, the spread of the stream will result in metal falling outside the mould. On the other hand, if the nozzle is too close to the top of the mould, the solidifying cone formed will bridge the gap, actually fusing to the sinkhead.

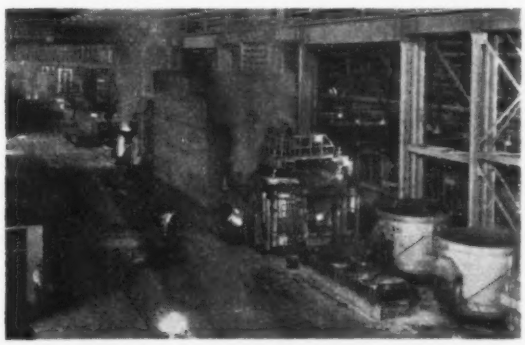


**Fig. 4—Electric furnaces**

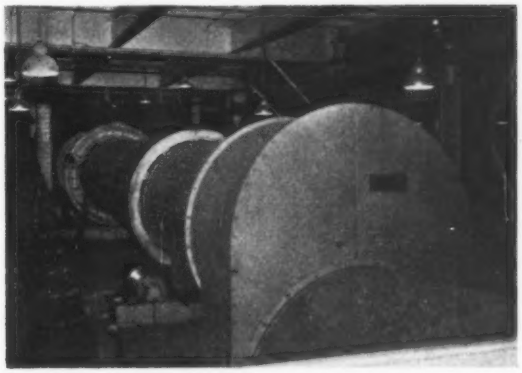


**Fig. 5—Ore dryer**

## 250-TON PRODUCTION UNITS

In addition to the 7-ton unit, two large 250-ton vacuum-degassing units are in operation in the electric-furnace melting department. A brief survey of the equipment in this shop, for servicing these large vacuum units, may be of interest.

### FURNACES

The electric furnace melting shop houses five conventional basic-arc furnaces, one rated at 28 tons and four rated at 50 tons. The furnace bottoms are lined with magnesite brick covered with a 6½-in. layer of rammed magnesite. The roofs are constructed of high-alumina brick and fitted for conventional 20-in. dia. graphite electrodes. Each furnace is equipped with automatic regulators for controlling power input from 10 000 and 18 500 kVA transformers. A general view of the melting units is shown in Fig. 4.

#### Drying of raw materials

For many years standard practice at Bethlehem has included the drying of ore, lime, alloys, and fluxes added to electric-furnace heats. This practice was set up for driving off moisture and thus eliminating a potential source of hydrogen from the raw materials. The effectiveness of the drying practice is reflected in the 3·4 p.p.m. average hydrogen content on tap as experienced over a period of several years.

The equipment used in the handling and drying of these materials is quite modern. Storage bins and conveyor belts service a 60-ft rotary kiln that heats ore, fluorspar, and sand to 300° F. Lime is dried in a separate rotary drier capable of heating to 1 500° F. Ferro-alloys are dried in charging boxes placed in a hearth furnace operating at 1 400° F. The three drying units are shown in Figs. 5, 6, and 7.

After bulk materials have been dried, they are stored in bins designed for easy discharging and weighing, as shown in Fig. 8.

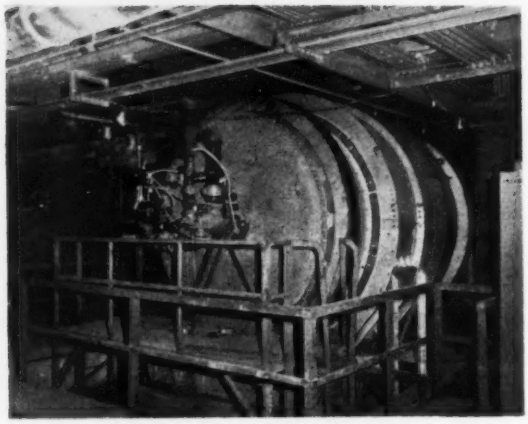


Fig. 6—Single-purpose lime dryer

**Melting and refining**

Standard melting practices have been established for all grades of steel made in the electric furnaces at Bethlehem. The practices are a carefully outlined step-by-step procedure from tap to tap and strict adherence is maintained throughout production.

All steels for heavy forgings are made with a double-slag practice: the first, or basic oxidizing slag, consists of lime and spar; the second, or reducing slag, is a mixture of lime, sand, spar, graphite, and silicon carbide in an approximate ratio of 30 : 3 : 2 : 2 : 4 respectively. Deoxidation of the bath is accomplished mainly through use of ferromanganese after slag-off and ferrosilicon before tapping. No aluminium is used for deoxidation in producing steel for forging ingots.

Temperature control is of paramount importance throughout all stages of melting. Pre-slag-off temperatures are in the range of 2 950–2 980° F, and pre-tap temperatures range from 2 875° to 2 920° F, depending on the grade of steel. All furnaces are equipped with automatic temperature-recording devices, and immersion hot-metal temperatures are obtained from furnace ladle, pony ladle, and ingot mould.

The propitious use of the five relatively small electric furnaces permits exceptional versatility for producing single ingots up to 120 in. dia. and weights of 500 000 lb maximum. In the production of multiple-heat ingots, there are naturally problems of critical scheduling and timing. However, the magnitude of these problems has been minimized by the years of experience in production. The problems presented from this standpoint are more than offset by the advantages attainable in the production of large heavy ingots of relatively uniform composition. Minimizing of segregation in the ingot is accomplished by controlling the analysis according to the tapping sequence of the various heats. As an example, in the production of a 120-in. ingot which requires a five-heat combination, carbon in the first or bottom heat will be set at 0·25%, the second heat at 0·23%, the third heat at 0·21%, the fourth heat at 0·20%, and the fifth heat or the sinkhead at 0·18%. Decreasing the carbon content in the successive heats in this way offsets the tendency for excessive carbon segregation in the top of the ingot.

**Pouring pit**

The pits of the electric furnace melting department were completely redesigned and rebuilt to accommodate the vacuum-degassing facilities. Dome-type degassing tanks were installed because of limited overhead clearance in the existing buildings. As a necessary part of the layout, portable pouring platforms were provided as well as adequate lay-down and service areas for domes, moulds, ingots, stools, and sinkheads. Complete facilities were already available for preheating furnace ladles and pony ladles and for drying stopper rods and nozzles.

**Ladles**

Furnace or tapping ladles are of conventional welded construction and made within the Bethlehem plant. Their capacities are rated at 50 and 76 tons. Ladle linings are of high-quality alumina fireclay brick. All newly lined ladles are dried with coke-oven gas for 8 h before use, and before each tap ladles are kept hot with an open gas flame.

Intermediate or pony ladles are used to transfer metal from the furnace ladle to the mould in the vacuum tank. These relatively small ladles, of 12½ tons, are necessary to maintain a continuous pour during transfer of the respective furnace ladles in pouring multiple heat ingots. The ladles are equipped with two nozzles to avoid overflowing of the pony ladle in the event of the loss of a stopper rod in the furnace ladle. The same drying precautions are practised on pony ladles as on furnace ladles. Before use, ladles are cleaned of dirt and refractory particles.

**Nozzles**

Refractory nozzles of various sizes are used in the furnace and pony ladles. All nozzles are saturated in tar at 260° F for 72 h, dried at 460° F for 72 h, and then wire-brushed and ground on the stopper-rod seat. This procedure has resulted in the assurance of good shut-offs and freedom from leaking nozzles.

**Moulds and sinkheads**

The standard moulds for heavy forging ingots are round corrugated moulds with 24 corrugations. Moulds up to and including the 120-in. size are of the conventional one-piece design. The 134-in. corrugated mould, used for certain carbon-steel or low-alloy ingots with weights up to 700 000 lb, is of a three-piece

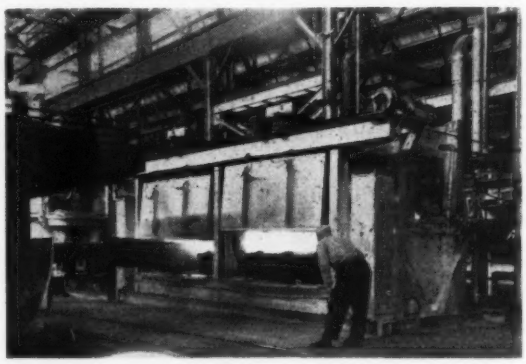


Fig. 7—Alloy-drying furnace

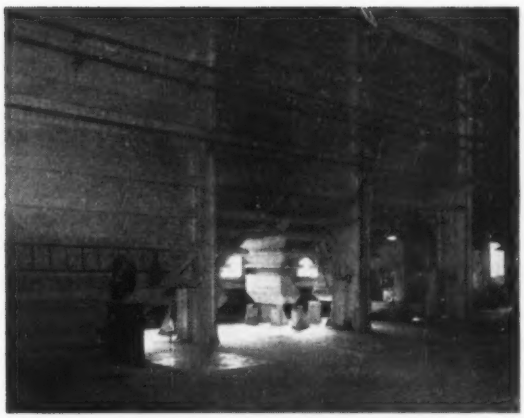


**Fig. 8—Storage bins for dried materials**

design. They also are corrugated, having 24 corrugations.

Mould design allows the downsetting of sinkheads into the mould to permit maximum flexibility of ingot weights from a given mould. In setting up for pouring, sinkheads are anchored to the mould to prevent rising during the pour. However, separate holding pins are released after completion of the pour to allow the sinkhead to settle with shrinkage, thus preventing the development of hanger cracks.

### Pit practice

Standard practice includes a close inspection of all ladles and nozzles, as well as moulds, upon completion of set-up and before tapping. Nozzles are checked for proper size, and each mould and sinkhead is checked for proper height. In producing ingots from multiple heats, a time schedule is set for each furnace to ensure sufficient time in the ladle for proper clean-up without loss of temperature for each heat. Metal temperatures for tapping are required in each furnace and, of course, pouring temperatures throughout the pouring of the ingot. These temperatures include pony-ladle temperatures for each furnace ladle as well as a temperature taken at the seal when the metal reaches the base of the sinkhead. A typical series of temperatures would record the tap at 2 910° F, pony ladle temperature of 2 820°, with a seal temperature in the ingot of 2 760°. All heats are sampled for the usual chemical analysis and hydrogen determination.

A prescribed holding time in the mould before stripping is part of standard practice. This time, in hours, is approximately half the diameter of the ingot. About 4 h before stripping, the sinkhead flask is removed to permit the lifting lugs to cool for subsequent stripping of the ingot from the mould.

### Capacity

The two 250-ton units are identical and capable of handling ingots up to 120 in. in diameter weighing 500 000 lb. The first large unit was put in operation in July, 1957, and the second unit in January, 1958. The number of ingots that can be produced in continuous operation of the two units is a variable, depending upon the size of the ingots produced. Under normal operating conditions and an average order book, from the standpoint of size, production will average 26–28 ingots per month.

Figure 9 shows the two large units in the electric

**Fig. 9—Electric furnace melting department 250-ton vacuum-degassing tanks**

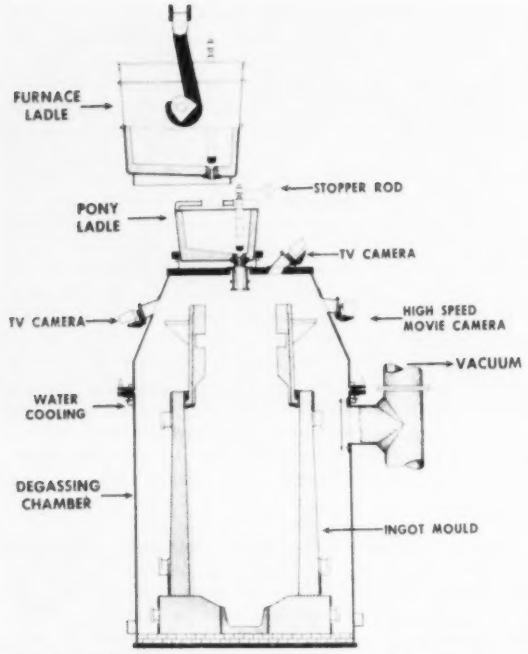



**Fig. 10—Schematic diagram of 250-ton vacuum-degassing unit**

furnace shop, indicating one unit in operation during the pouring of an ingot. The second unit appears with the domed head removed and a 92-in. mould and sinkhead set up in the tank.

Except for the larger size and the use of a 4-stage steam-ejector vacuum system, the 250-ton units are quite similar in design and construction to the 7-ton development unit.

Figure 10 shows schematically the general arrangement of the large unit. As indicated, the upper ladle is the furnace ladle, into which the steel is tapped from the furnace. Directly under this ladle, and seated on the top of the vacuum tank, is a small ladle designated as a pony ladle. There is a vacuum-tight seal between this ladle and the top of the tank on which the ladle is set. An aluminium rupture disc is attached to the pony ladle directly under the pouring nozzle; this seal is obviously necessary to obtain the vacuum in the tank before pouring. The mould is set up inside the vacuum tank and centred directly under the rupture disc. The high-speed motion-picture camera and the two television camera locations are also indicated.

### Vacuum tank

The vacuum tank proper consists of two sections, a stationary straight-sided bottom portion, having an inside diameter of 15 ft 6¼ in., with a height of 19 ft 2 in., and a removable cover or dome. The dome decreases to 10 ft 8 in. i.d. at the top, having a height of 8 ft 7½ in., resulting in an overall inside height of 27 ft 9½ in. These sections are constructed of 1½-in. thick plate, except that a 6-in. thick plate is used for the closure in the top of the dome. A Neoprene O ring is used to provide a vacuum seal between the dome and the stationary section of the tank. An additional Neoprene O ring provides a seal between the pony ladle and the top of the closure head. Both joints are water-cooled to protect the rings. In addition, the bottom portion of the tank is equipped with a circular perforated water-cooling ring on the outside of the tank.

### Vacuum system and instrumentation

The vacuum for the tank is produced by a 4-stage steam-jet ejector system designed to operate with a minimum steam pressure of 135 lb of 450° F steam. The system can evacuate 318 lb per hour of 19·3 molecular weight gas at 212° F and an absolute pressure of 1 mm Hg. The first-stage steam-ejector pump is non-condensing. Barometric condensers are located between the second and third and between the third and fourth stages to reduce the volume of steam and gas being handled through the succeeding stage. The fourth and final stage exhausts to the atmosphere.

The steam-ejector system was selected as the preferred method of producing vacuum mainly because of lower capital investment and lower maintenance costs. The decision to select this system has been vindicated in every respect of operation including low capital investment, low installation cost, minimum maintenance, and satisfactory degassing capacity to maintain vacua consistently at the desired level during pouring.

The arrangement of the jets, control valves, and tanks is shown in Fig. 11.

The design specifications covering the steam-jet system are as follows:

| | |
|---|---|
| Operating pressure, μ Hg abs. | 1 000 |
| Air capacity, lb/h | 318 |
| Steam pressure, lb/in² gauge | 135 |
| Steam consumption, total lb/h | 8 900 |
| First stage | 970 |
| Second stage | 4 395 |
| Third stage | 1 245 |
| Fourth stage | 2 290 |
| Condenser water consumption, total gal/min | 850 |
| Condenser water temperature—in, ° F | 90 |
| Condenser water temperature—out, ° F | 107 |
| Actual steam consumed, lb/h | 9 700 |

In starting up the ejector system to obtain the desired vacuum, steam is first turned on to the fourth stage, then successively to the third, second, and first stages as each stage comes up to maximum performance. When the last three stages are operating effectively, the pressure in the tank will be at about 3–5 mm Hg. Introduction of the first stage then reduces the pressure to a range of 300–400 μ, at which time the unit is ready for vacuum pouring of the steel.

The units are equipped with two television cameras, one directed toward the pouring nozzle and the other aimed into the mould, so that the condition of the stream can be followed at all times. The high-speed motion picture camera provides a practical means of recording the result of any practice changes, in connection with pouring or set-up, for subsequent study and evaluation.

The vacuum pressures are measured by gauges, which operate on the principle that the conductivity

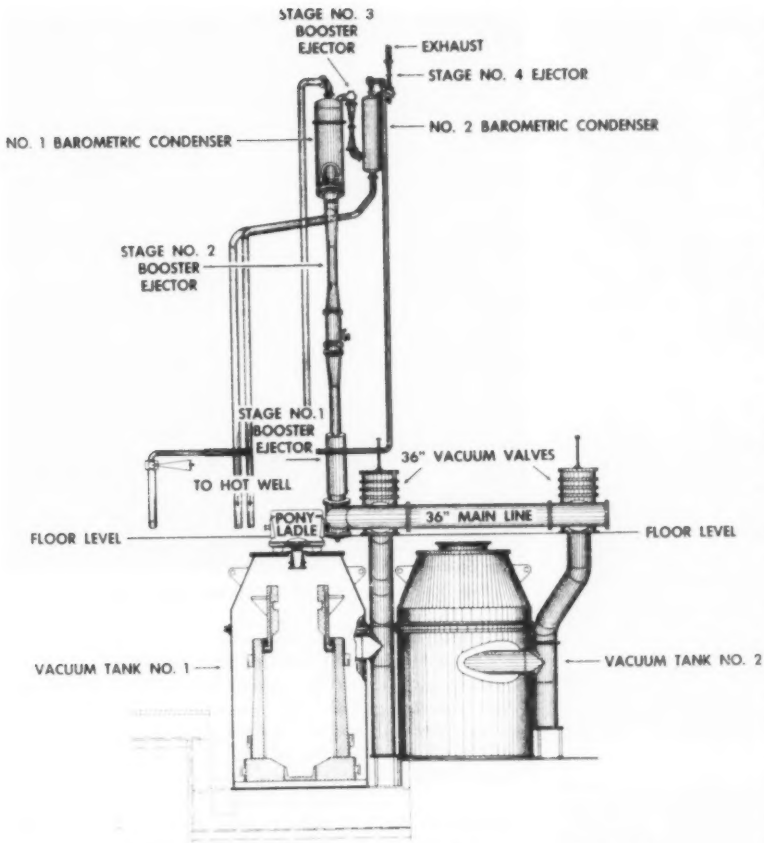



Fig. 11—Steam-jet system for 250-ton vacuum tanks

of ionized gas will change according to the quantity of the gas present.

Figure 12 shows the panel board and instrumentation for the control of the operation. It is conveniently located adjacent to the pouring platform. The control board includes the necessary recorders and gauges for indicating metal temperature during the pouring, vacuum pressure, and water and steam consumption. Controls for breaking vacuum, as well as for admitting inert gas to purge the tank after the pour, are also located here. The television screens for the two cameras are located on the board and can be seen in Fig. 13 more clearly.

The upper screen shows the steel emerging from the nozzle and the immediate breaking up of the stream as a result of the vacuum. The lower screen, directed at the top of the mould, gives a further idea of the breaking up of the stream as it enters the mould.

**Preparation of tank and mould**

Before the pumping-down operations, all oxide and dust must be removed from the tank, the main line, and the main vacuum valve to ensure effective operation of the unit. In addition, all moisture must be removed by thoroughly drying the mould and sink-head. The ingot mould is cleaned free of all oxide, heated to remove any moisture, and then placed on the stool in the tank. The sinkhead is prepared separately and is dried for 24 h above 1 000° F. The refractory lining of the sinkhead must be of high-quality alumina brick and cement to withstand a minimum temperature of 3 000° F and must be capable of resisting the erosion from the molten metal stream. The packing material, used between the sinkhead and the mould,

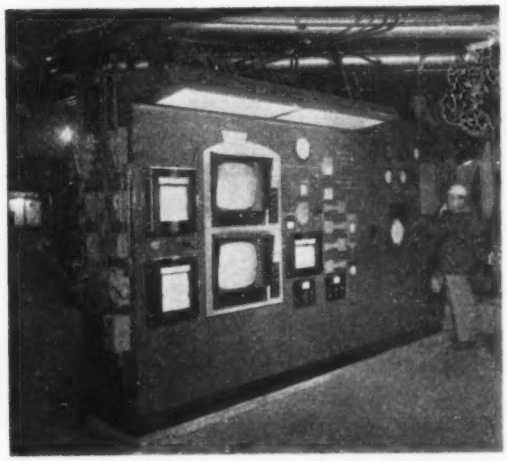


Fig. 12—Control panel for vacuum-degassing units

Fig. 13—Television monitors on control panel

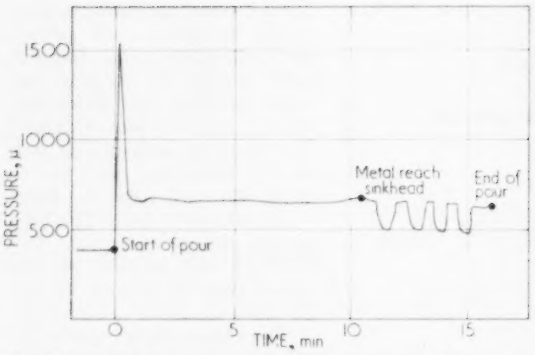



**Fig. 14—Typical pressure curve during pour of ingot in 250-ton vacuum-degassing unit**

must have good ramming properties to seal the sinkhead junction effectively and prevent metal from cutting through the seal. This packing is followed by a refractory which will withstand the high temperature and the pull of the vacuum. After the packing is completed, the entire assembly is again heated for several hours to remove moisture. While the need of a good seal at the sinkhead and mould juncture is important in the air casting of ingots, it becomes even more important in vacuum pouring, since any metal leak will result in metal spilling into the tank. As a final step, the mould is siphoned to remove loose particles of refractories or dust before the dome and pony ladle are placed in position on the tank.

**Operation of vacuum unit**

In operating the vacuum unit, the desired blank-off pressure is obtained in the sequence previously described and then the steel is poured from the furnace ladle into the pony ladle. When sufficient metal has been poured into the pony ladle to ensure a continuous pour of the ingot, the stopper rod of the pony ladle is opened and the aluminium seal broken by the hot metal stream.

Figure 14 indicates a typical log of the pressure during these operations and during the pour of an ingot, which in this case happens to be a 92-in. dia. ingot. An initial surge of pressure occurs when the metal is first exposed to the vacuum, after which the pressure drops and remains fairly constant during the pouring of the body of the ingot and until the seal is reached. When the metal reaches the seal at the base of the sinkhead, which is indicated by the temperature rise on a thermocouple located at the seal, pouring is interrupted and the pressure drops momentarily. It then reaches its previous level when pouring is resumed. The series of fluctuations between the point where the metal reaches the seal and the end of pouring are a result of intermittent opening and closing of the pony-ladle nozzle during the pour of the sinkhead.

Caution must be exercised in pouring the sinkhead because of the rolling action of the metal resulting from its exposure to the low pressure. If the rolling becomes too active, metal may spill over the sinkhead into the tank, and to prevent this it may be necessary to shut down some stages of the vacuum system.

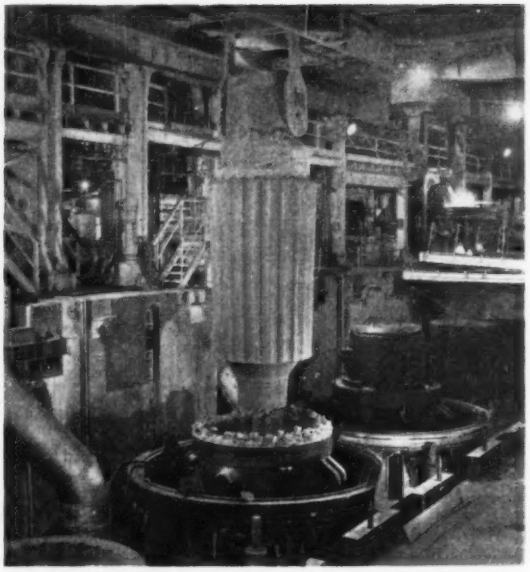


**Fig. 15—Stripping of 92-in. vacuum-cast ingot**

When pouring is completed, the vacuum ejectors are shut down in the order of the first, second, third, and fourth sequence, the opposite of the start-up sequence.

To guard against the possibility of explosion, resulting from exposure of inflammable gases or dust in the tank to air when the seal is broken, the tank is purged with an inert gas, after which the vacuum-break valve is opened. After atmospheric pressure is reached in the tank, the various hose connections are

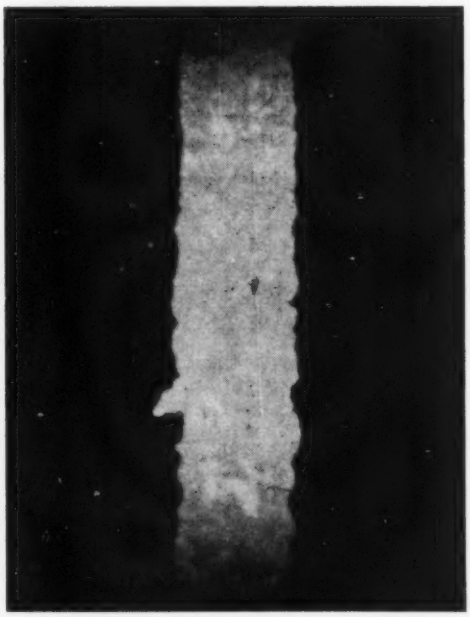


**Fig. 16—Metal stream during pour of conventionally cast ingot**

removed and the pony ladle is removed from the tank. At this point, a test sample of molten steel for hydrogen determination is obtained from the sinkhead. When compared with the test samples removed from the pony ladle, this permits an evaluation of the effectiveness of the vacuum pour in removing hydrogen from the steel.

The platform and dome must be removed quickly to avoid excessive heating. The ingot is allowed to remain in the tank the necessary time for solidification, which, in the case of the 92-in. ingot, is 40 h. During this period the outside of the tank is water-cooled through a circular perforated pipe located just below the main flange.

Figure 15 shows the ingot, weighing approximately 280 000 lb, being stripped from the mould at the conclusion of its solidification period. The loose brick, apparent on the top of the mould, is the result of removing the sinkhead flask before stripping the ingot from the mould. In the case of light-weight ingots, both the mould and the ingot may be removed from the tank before stripping of the ingot. Shop practice in this respect is determined on the basis of capacity of cranes available for performing the operation.

**Pouring stream**

The efforts on the part of the melting shop to maintain a tight, rope-like stream in the pouring of an ingot is familiar to all operators. Such a stream is shown in Fig. 16.

It is well recognized that a stream of this type in conventional air casting minimizes the opportunity for the metal to oxidize and thereby furthers the cause of cleanliness in the ingot. Inclusions in the finished piece have been encountered too often and have been attributable to a spreading stream, as the result of a faulty ladle nozzle.

In contrast to the desired rope-like stream in air casting, Fig. 17 shows the stream from the nozzle when casting *in vacuo*. The stream of hot metal is literally torn into relatively small droplets on entering the vacuum tank. As previously mentioned, the stream assumes a wide-angled conical shape during the drop into the ingot mould.

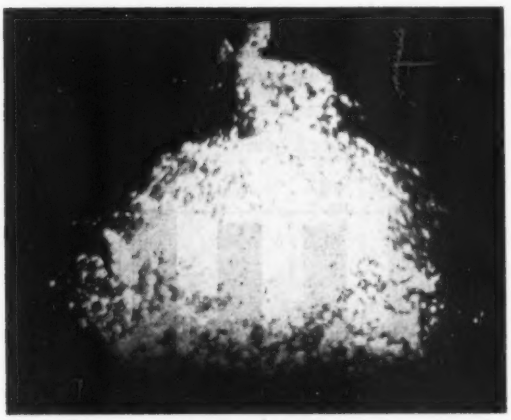


**Fig. 17—Metal stream during pour of vacuum-cast ingot**

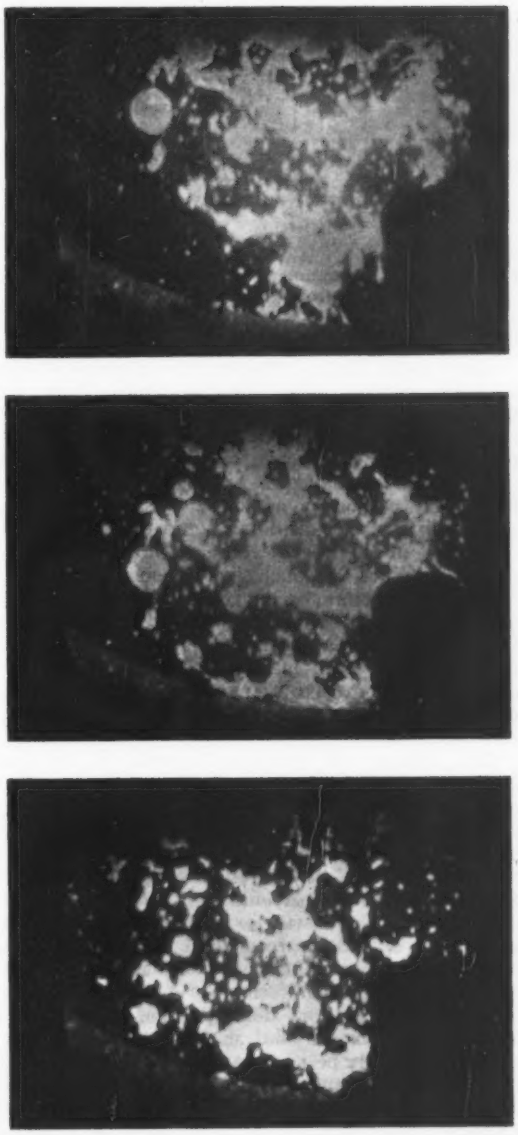


**Fig. 18—Breaking-up of metal stream during vacuum pouring**

The effectiveness of the vacuum in breaking up the stream and what happens to the individual globules or droplets is further indicated in Fig. 18. The photograph represents three individual frames from a reel of film taken with a high-speed moving-picture camera. The single globule apparent in the upper frame is shown collapsing in the second frame and finally disperses into numerous small droplets as shown in the last frame. There can be little question as to the effectiveness of the vacuum in exposure of the stream for maximum hydrogen removal; this is in sharp contrast to the conventional stream shown in Fig. 16.

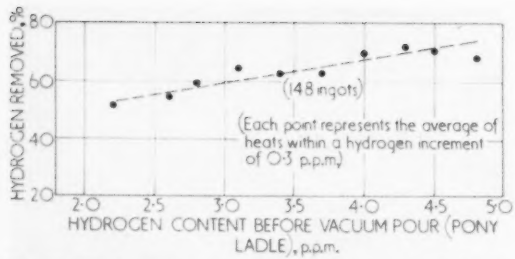



**Fig. 19—Hydrogen removal based on pony ladle hydrogen (all grades)**

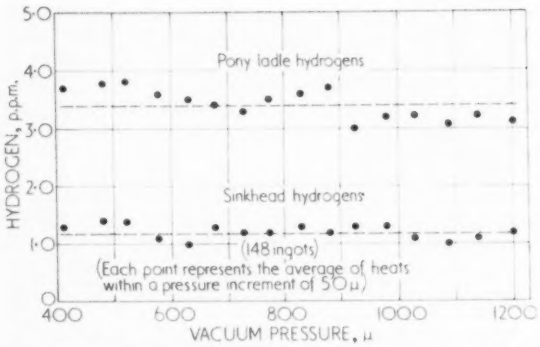



**Fig. 20—Hydrogen versus vacuum pressure (all grades)**

## HYDROGEN REMOVAL

Gases other than hydrogen are extracted during vacuum pouring. Gas samples taken from the vacuum line indicate on analysis that the gases coming out of the tank consist of approximately equal parts of hydrogen, nitrogen, and carbon monoxide. A typical analysis indicates hydrogen 31%, nitrogen 36%, and carbon monoxide 29%; the remaining 4% consists of carbon dioxide and oxygen. Whilst reduction of hydrogen content in the vacuum-poured product is conclusively demonstrated, the effect of the vacuum on oxygen and nitrogen content in the product is less clear. Published information on this matter is quite at variance. Since interest at Bethlehem has been mainly in hydrogen, we have concentrated on this element and have not yet obtained sufficient data on other gases to warrant factual comment.

The amount of hydrogen removed in vacuum degassing in the 250-ton unit is similar to that experienced in the 7-ton unit. Samples taken from the pony ladle compared with samples taken from molten metal in the sinkhead of the degassed ingot indicate that an average of 63% of the hydrogen is removed. Figure 19 shows the percentage of hydrogen removal and indicates the scatter experienced. In this connection, it should be recognized that any specific hydrogen figure should be considered with due cognizance of the limitations of present analytical methods for hydrogen determination. Present-day accuracy in hydrogen analysis for controlled production can probably best be set at $\pm \frac{1}{2}$ p.p.m. Whilst a given laboratory can reproduce results with reasonable accuracy, the reproducibility between different laboratories is less exact.

In the production of vacuum-degassed ingots it is only natural that the degree of vacuum should be regarded as one of the most pertinent factors. In the evaluation of this factor from the standpoint of its effect on hydrogen removal, a consistent and standard practice must be maintained in the melting shop. It is on this basis that the following data are presented, illustrating the effect of the vacuum pressure on the hydrogen content of the ingot, as indicated by molten metal samples from the sinkhead.

Figure 20 is prepared on the basis of data on 148 ingots. The hydrogen content has been averaged for heats within each 50-$\mu$ increment. As indicated by the upper line, the hydrogen content of the metal in the pony ladle is reasonably constant, reflecting the standard melting practice. The lower line indicates the hydrogen content determined on ingots poured *in vacuo* between somewhat under 500 $\mu$ and up to 1 200 $\mu$. The relatively flat and straight line is clearly indicative of the effective hydrogen removal in pouring anywhere within this vacuum range. Obviously, considerable latitude is permissible for selecting satisfactory operating conditions. The establishment of this fact, ensuring a satisfactory end product, is of importance by virtue of the latitude permitted for the further study of the effect of vacuum level on other phases, such as ingot surface and cone formation during pouring.

Experience has also shown that the degree of hydrogen removal remains fairly constant over a wide range of pouring rates.

The various compositions used for heavy forgings have been evaluated on the basis of their possible effect on the percentage of hydrogen removed during vacuum pouring. The result of this analysis is shown in Fig. 21. It is apparent that there is only an insignificant difference in the amount of hydrogen removed for the various compositions.

## HYDROGEN IN THE FORGING

Figure 22 shows the hydrogen content throughout the processing of a large number of forgings. The graph shows a comparison indicating the average hydrogen content of steel from the pony ladle and of molten metal from the sinkhead of the vacuum-cast ingot. It also shows the hydrogen as determined in the actual forging from both a conventionally cast ingot and from a vacuum-cast ingot at locations

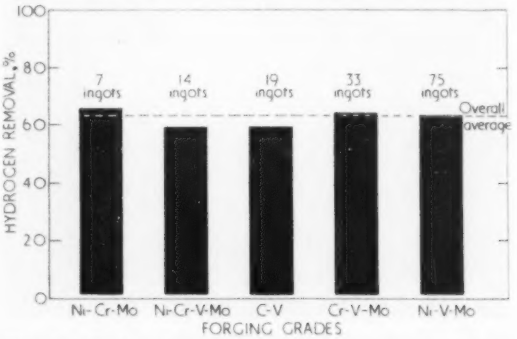
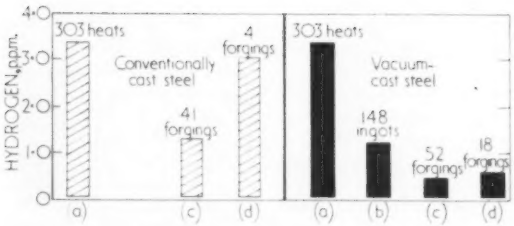



**Fig. 21—Hydrogen removal for various compositions**




(*a*) Hydrogen before pour
(*b*) Hydrogen after vacuum pouring (sinkhead)
(*c*) Hydrogen 6 in. below surface of forging after treatment
(*d*) Hydrogen at centre of forging after treatment

**Fig. 22—Average hydrogen content during processing of forgings from conventional and vacuum cast ingots**

6 in. below the surface and at the centre of the forging.

The average hydrogen content before pouring (i.e. as taken from the pony ladle) averages 3·4 p.p.m. While the chart does not show an average figure from the sinkhead of air-cast ingots, it is obvious that the figure would be equivalent to, or slightly higher than, the pony-ladle hydrogen, depending upon the humidity at the time of casting. In contrast to this, it is gratifying to note that an average hydrogen of 1·2 p.p.m. is obtained from the sinkhead on the vacuum-poured ingots.

Subsequent processing of the ingot, involving heating for forging, preliminary or conditioning treatment, and heat-treating for mechanical properties, results in hydrogen diffusion and a further reduction in hydrogen content of the piece. In evaluating forgings as to hydrogen content, a point 6 in. below the surface of the forging is nominally the greatest depth that can be explored; this is the inner end of a radial test bar. In contrast to slightly less than 1·5 p.p.m. hydrogen on air-cast material, on vacuum-cast material an average of 0·5 p.p.m. is obtained.

An even more impressive contrast between these two types of material is shown by the hydrogen content at the dead centre of the forging. Whereas

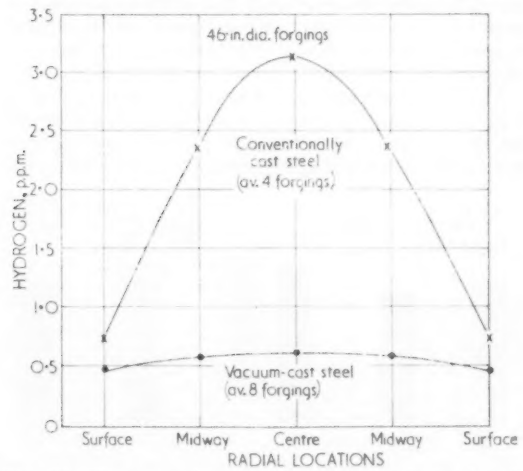



**Fig. 23—Cross-sectional distribution of hydrogen in Ni–V–Mo forgings from conventional and vacuum-cast ingots**

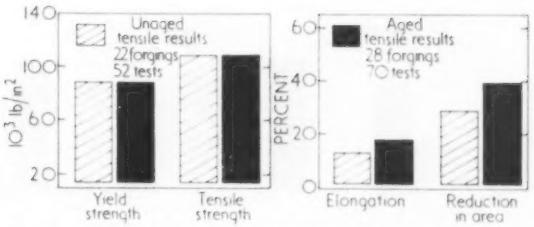



**Fig. 24—Average transverse mechanical properties on Ni–V–Mo forgings from conventionally cast ingots**

forgings from conventionally air-cast ingots have an average hydrogen content at the centre practically the same as that in the pony ladle or the sinkhead of the ingot, in forgings produced from vacuum-cast ingots the hydrogen content at the centre is comparable with the hydrogen at a point 6 in. below the surface. The number of ingots or forgings examined has been materially increased since the preparation of these data, but the average figures indicated are not affected by the additional data.

Figure 23 shows hydrogen distribution throughout the cross-section of 46-in. dia. forgings from four conventionally cast ingots and eight vacuum-cast ingots, all of Ni–V–Mo steel. The difference in distribution of hydrogen content throughout the forgings on the two types of material is obvious. The uniformity throughout the cross-section and low level of the hydrogen content on forgings from vacuum-cast steel is indeed gratifying to both the producer and the ultimate forging consumer.

## MECHANICAL PROPERTIES

It has been generally recognized that some steels are sensitive to hydrogen embrittlement, as indicated by their low ductility on tensile testing. Ageing of tensile bars by heating to temperatures as low as 212–500° F, materially below the last tempering temperatures will result in improved ductility on the aged tests as contrasted to unaged tests. This effect is indicated in Fig. 24, which represents Ni–V–Mo forgings from conventionally air-cast ingots. Whilst there is little or no change in tensile and yield strengths on aged or unaged tests, an improvement is indicated in elongation and reduction in area on the aged tests.

A contrasting picture on Ni–V–Mo forgings from vacuum-cast ingots is shown by Fig. 25. Vacuum pouring has eliminated the difference between aged and unaged testing. Not only are the tensile and yield strengths the same on aged and unaged testing, but also both the elongation and reduction in area are practically the same in both types of test, which is most significant.

The two previous charts have contrasted the difference between aged and unaged test bars on the two types of material. Since our study has indicated the product of vacuum-cast material to have equally as good properties, whether measured by either aged or unaged test bars, the next comparison should logically be between tests on this product and tests on forgings from conventionally cast material in both the aged and unaged condition.

In Fig. 26 the tensile and yield strength have been omitted as they have been found to be comparable,

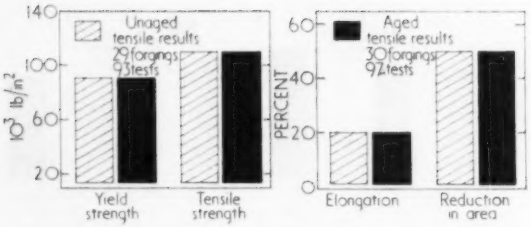



**Fig. 25—Average transverse mechanical properties on Ni–V–Mo forgings from vacuum-cast ingots**

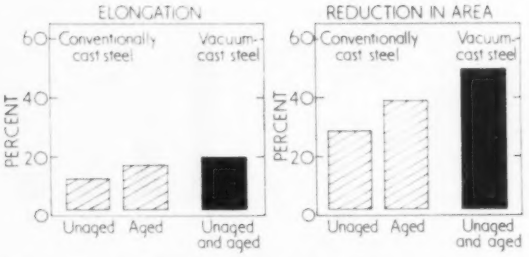



**Fig. 26—Unaged versus aged ductility properties on Ni–V–Mo forgings from conventional and vacuum-cast ingots**

and also unchanged by ageing. The ductility characteristics alone are compared. The improvement in ductility, both in elongation and reduction in area, on the vacuum-cast product is indeed obvious and is one of the important benefits of this process to both producer and consumer. The graph clearly indicates the superior ductility of the vacuum-cast product over the conventionally cast product, even when compared with the aged test results.

In studying Ni–Cr–V–Mo and Cr–V–Mo grades, it is recognized that neither of these grades is sensitive to the type of hydrogen embrittlement commonly indicated by a lowering of the tensile ductility. The properties obtained by aged and unaged testing are comparable and are omitted in the presentation of these grades. The data submitted for comparison of the two types of material are all on aged tests. However, here again the vacuum-poured product shows an improvement in ductility, although possibly to a slightly lesser degree than in the case of Ni–V–Mo. This is true on both the Ni–Cr–V–Mo grade and the Cr–V–Mo grade forgings.

Figure 27 indicates the average properties of Ni–Cr–V–Mo forgings produced from conventionally air-cast ingots and also from vacuum-cast ingots. At equivalent, or even slightly higher, yield strengths and tensile strengths, there is an improvement of the vacuum-cast product in both elongation and reduction in area.

In the case of Cr–V–Mo forgings, essentially the same condition is found to exist. Figure 28 shows that on equivalent-strength material, the vacuum-cast product shows a higher ductility both in elongation and reduction in area.

The improvement in ductility of the vacuum-cast product as compared to the conventionally cast product for Ni–Cr–V–Mo and Cr–V–Mo grades is obviously a welcome feature both to producer and consumer. Since hydrogen has been indicated not to affect materially the properties of these two grades of forgings, it is evident that vacuum pouring has produced benefits to the finished product other than those attributable to low hydrogen content.

## CLEANLINESS

In attempting to arrive at an explanation for the improvement in ductility on the vacuum-cast product, it is only natural to explore the two types of material for relative cleanliness. To accomplish this, the tensile bar fractures were examined under magnifying binoculars. On the basis of the inclusions present, they were rated on a numerical system in which the sizes of the non-metallic inclusions have been arbitrarily graded from 0 to 3·0. As a result of this study, the average rating for the conventionally cast product for all grades was found to be 1·7, compared with an average rating for the vacuum-cast product of 0·2.

The ratings for the three grades of material, Ni–V–Mo, Ni–Cr–V–Mo, and Cr–V–Mo, and the average of all the ratings when plotted in the form of distribution curves, are shown in Fig. 29.

The results of this study present an interesting picture. It is indicated on the conventionally cast product, where the average rating was 1·7, that the inclusion ratings are distributed over a broad range from 0 to 3·0. On the other hand, the inclusion rating on forgings from the vacuum-cast product averages 0·2. This is not only a gratifyingly lower average but it is also found to be distributed over a very narrow range of 0 to 1·0.

Whilst the microscope is not considered as a reliable tool in evaluating the cleanliness of forgings, it is of material value from the standpoint of study and identification of individual inclusions. A microscopic study was therefore made as a supplement to the fracture findings for comparison of the two types of product. Results of this study indicate that silicate-type inclusions are fewer and materially smaller in the vacuum-cast product. Other types of inclusions appear to be similar in the products of the two processes.

The effect of vacuum pouring on segregation is of material interest, particularly as regards the sulphide streaks or 'ghost lines' found in heavy forgings, and normal in the solidification of large masses. Although

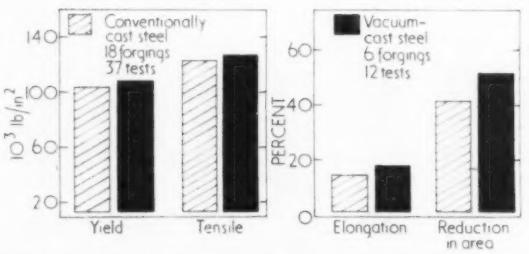



**Fig. 27—Average transverse mechanical properties on Ni–Cr–V–Mo forgings from conventional and vacuum-cast ingots**

information on this subject is limited, experience in the examination of certain forgings where design necessitates appreciable machining of the as-forged diameter has developed some information. In such instances, where appreciable amounts of metal have been removed, inspection has indicated that segregation of this type, as found in the vacuum-poured product, is similar to that found in a conventionally poured product.

Macro-etch and sulphur-print testing of complete cross-sections of forgings have naturally been quite limited. However, based on the work performed to date, our findings have confirmed the result of visual inspection of machined surfaces. There appears to be little or no difference in the product of the two types of material. In this connection, it is probably only natural to anticipate that such segregations, being the result of the natural laws of solidification of large masses, would not be materially affected. Further work will have to be done to determine whether the intensity and dispersion of these segregates are of any lesser degree in the product of the vacuum-poured ingot.

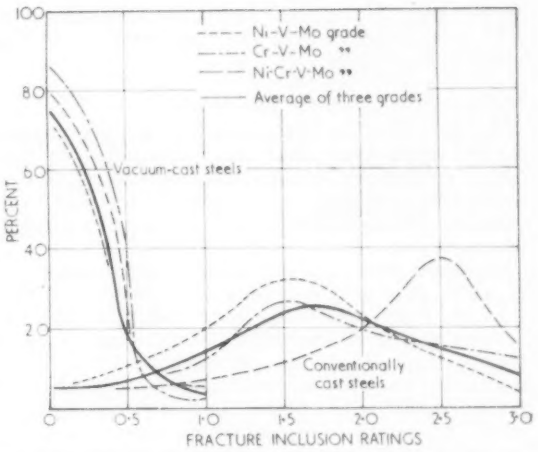



**Fig. 29—Distribution of fracture inclusion ratings on forgings of three grades from conventional and vacuum-cast ingots**

## OTHER PROPERTIES

A summary of comparison of other properties in the products of air-cast material and vacuum-poured material indicates the following:

(1) *Transition temperature*—In recent years a desire to avoid brittle failures, particularly in rotor forgings, has resulted in an emphasis on obtaining low transition temperatures. Charpy V-notch specimens have generally been accepted for determining this property and the temperature at which the 50% granular fracture is obtained is defined as a transition temperature. A comparison of the results indicates that the vacuum-cast product is directly comparable with the conventionally cast product.

(2) *Impact properties*—A study of results obtained to date indicates that the impact properties on the product from the two types of material are comparable.

(3) *Grain size*—A comparison of grain size on the basis of McQuaid–Ehn tests shows no difference between vacuum-cast and conventionally cast material.

(4) *Magnetic permeability*—A comparison of limited test results up to the present time does not indicate any difference in magnetic permeability when comparing the two types of material.

(5) *Stress-rupture*—Stress-rupture tests have not progressed to the extent that any conclusions can be reached at this time. However, it is not anticipated that there will be any difference between vacuum-cast and air-cast material.

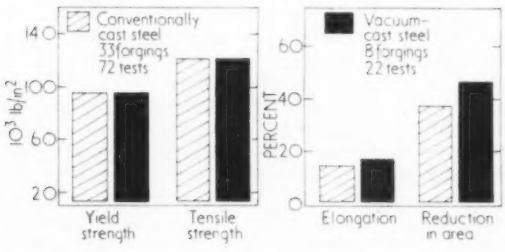



**Fig. 28—Average transverse mechanical properties on Cr–V–Mo forgings from conventional and vacuum-cast ingots**

## SUSCEPTIBILITY TO FLAKING

The benefits and advantages of vacuum degassing that have been discussed are all important but perhaps of most importance, and justifying particular emphasis, is the improvement in internal soundness of large forgings from the standpoint of 'flaking' or micro-cracks. The product of the degassed ingot, with its lower hydrogen content, has proven to be greatly less flake-sensitive as determined by sonic results on over 200 large forgings, all of which have failed to reveal a single instance of 'flaking.' This is a materially different experience from that commonly found with the product from conventionally cast material.

To prove this characteristic, numerous forgings of 15–60 in. dia. from discard material have been variously handled from the forging press. Methods of handling have varied from air cooling from the forging heat to greatly reduced or abbreviated preliminary treatment cycles. Subsequent examination of these pieces by sonic testing has indicated freedom from flakes and has firmly established the virtue of vacuum degassing in the prevention of this type of difficulty.

The reduced susceptibility to flaking is regarded by the producer as welcome insurance. However, from the standpoint of treatment cycles, it must be remembered that the maximum and minimum temperatures used are essentially for the purpose of refinement in relation to the microstructure of the material. For this reason, it is readily conceivable that a treatment, as such, cannot be eliminated because of the use of vacuum-cast material. The possible modification of cooling rates can only be determined after exhaustive experimentation. Standard treatment practices are being maintained until an adequate background has been established. It should be recognized that air cooling of material from the

press is not a factor in the production of forgings but only a tool to measure the susceptibility of the material to 'flaking.'

## COST FACTORS

Ton per ton of ingots produced, the vacuum-poured ingot is more costly than the conventional air-cast ingot. Apart from the tremendous capital expenditure for the installation of equipment necessary for vacuum pouring there is also a definitely greater operating cost.

Assuming that present melting-shop personnel can be fully educated as competent operators of the vacuum equipment (experience indicates this is entirely practical), there must of necessity be additional personnel used in this process. The proper setting up of a mould in a tank, compared with ordinary practice, is a longer and more tedious operation. The man-hours per ton are therefore increased.

The refractories for the sinkhead, including packing materials, etc., must of necessity be a specialized product suitable for the conditions existing in actual production. In pouring a vacuum ingot with the desired spreading or spraying stream, the sinkhead refractories must resist the hot metal temperature and erosive effect of the continuous spraying of the stream throughout the entire pouring. Certain additional material costs are therefore inevitable.

Many of the hazards or calculated risks incident to the production of large ingots become considerably greater in the production of the ingot by the vacuum process. Large ingots are usually the product of multiple heats; the failure of one of the stopper rods or nozzles can make it impossible to control or shut off the flow of metal and can readily be conceived as presenting a serious operating problem. Equipment damage under such a condition is considerably greater in the case of vacuum pouring. The breaking out or bleeding of an ingot between the bottom of the mould and the stool or a break-out at the base of the sinkhead are indeed serious problems when they occur inside the vacuum tank. Spillage of metal over the top of the sinkhead, as a result of inability to shut off the flow, again is much more damaging when in the vacuum tank. Incidents such as these can result not only from faulty workmanship, but also as the result of material failure of the ceramics or packing used in setting up the mould. Carefully controlled practices and use of the best available materials result in infrequent occurrences of such incidents. When, on occasion, they are experienced, they are exceedingly expensive.

Fortunately, experience to date indicates very satisfactory ability of the equipment to maintain vacuums. However, it is only practical to recognize that any failure in this equipment will result in an ingot poured only partially *in vacuo*. Application of such an ingot may or may not be feasible on a production order and consequently may result in a scrapped ingot.

These comments are not intended to detract in any way from the advantages of the vacuum-poured product, but are intended only to emphasize some of the complexities incident on production by the process and logically reflect in a more costly product.

## FUTURE

Predicting the future is always hazardous. However, sound judgment does dictate the necessity of attempting to scan, within broad limits, the future potential for this process as it pertains to both the producer and the consumer.

An attempt has been made to present the benefits to be derived from vacuum degassing in the production of large ingots, as determined by experience to date. Naturally, there remains much work to be done in determining and evaluating other possible advantages from this process. Data available at this time as to the effect of vacuum pouring on characteristics such as creep, stress rupture, elevated-temperature properties, fatigue resistance, or magnetic characteristics are far too limited to reach any valid conclusions. Clarification of this picture can come only after exhaustive experimental and research work now in process.

In evaluating experience to date and in looking to the future, it should be recognized that vacuum pouring of large ingots has definitely been proved practical in every respect from an operating standpoint. The advantages resulting from the process are definitely worthwhile and can conceivably be considered essential for meeting the present-day high-quality requirements in large heavy forgings. In looking to the future, it is therefore only logical to say that the process is here to stay. It is a commendable, progressive step in the production of high-quality heavy forgings.

It seems likely that compositions prone to flaking and those subject to lowered ductility, by virtue of hydrogen embrittlement, will all be vacuum-poured in the production of large forgings intended for critically, highly stressed applications. The commonly used Ni–V–Mo grades appear as an outstanding example.

In regard to composition not subject to flaking and embrittlement, such as Cr–V–Mo, the vacuum-pouring process produces an improvement in cleanliness and a corresponding increase in ductility level. It is natural that the benefits from the standpoint of flaking would not appear as significant for this grade. However, the severe operating conditions to which Cr–V–Mo rotors are subjected have already placed this type of forging in the 'must class' for vacuum pouring.

It is a natural consequence that the important technical development of any process results in nearly everyone desiring that the benefit of this process should be reflected in the material used by him. Frequently this is true regardless of the requirements from the standpoint of serviceability. There are many applications for heavy forgings where service failure due to material deficiencies are practically unknown. The need for the vacuum-pouring process in the production of forgings of this type would appear to be questionable. However, highly critical inspection, particularly with the use of such tools as ultrasonic examination, can conceivably force this product into the category of the vacuum-pouring process. Clarification of this phase will undoubtedly come with the passage of time, further experience, and additional experimental and research work on the vacuum-poured product.

# *N E W S*

## Announcements and News of Science and Industry

## THE IRON AND STEEL INSTITUTE

### President-Elect

It was announced at the Autumn General Meeting on 10th December that Mr William Barr, O.B.E., Honorary Treasurer since 1953, has been nominated for election as President of the Institute for 1959–1960, to take office at the Annual General Meeting.

## NEWS OF MEMBERS

Monsieur **J. H. Alleaume** has joined Electricité de France, Clamart (Seine).

Mr **H. M. Bostandjis** has left the Metallurgical Equipment Export Co. Ltd.

Mr **W. Crafts** has been nominated Vice-President of the American Society for Metals for 1958–1959.

Dr **L. Douglas** has been appointed Assistant General Manager of South African Railways.

Sir **John Duncanson** has been elected Chairman of Metropole Industries Ltd.

Mr **R. S. El-Ansari** has joined the Fulmer Research Institute.

Mr **G. D. Elliot,** O.B.E., has been appointed deputy general works manager of the Appleby-Frodingham Steel Company; he will succeed Mr **J. D. Joy** as general works manager on 1st April.

Mr **G. F. Ewan** has been appointed General Manager (Production) of the Consett Iron Co. Ltd.

Dr **J. F. Ewing** has transferred to the Nuclear Facilities Plant of the Babcock and Wilcox Company, Lynchburg, Va., as Manager of Quality Control.

Mr **M. Farrow** has joined the Research and Development Branch of the U.K. Atomic Energy Authority, Culcheth.

Dr **G. G. Hatch** has left the Quebec Iron and Titanium Corporation to join W. S. Atkins and Associates, Toronto.

Mr **E. M. Holt** has been appointed General Manager of Hoy Carbides Ltd.

Mr **J. D. Joy** is to become deputy general manager of the Appleby-Frodingham Steel Company on 1st April.

Mr **J. A. Kilby** has been appointed a director of Colvilles Ltd.

Sir **Ernest Lever** is to retire from the chairmanship of Richard Thomas and Baldwins Ltd. in April. He will be succeeded by Mr G. C. R. Eley.

Mr **N. D. MacDonald** is now works manager (iron) at the Appleby-Frodingham Steel Company.

Mr **R. Marshall** has retired from the board of Colvilles Ltd., owing to ill-health.

Mr **B. J. Meadows** has left I.C.I. (Metals Division) Ltd. to take up the post of Lecturer at the College of Technology, Birmingham.

Mr **D. A. Oliver,** C.B.E., has been appointed a member of the Steering Committee of the Warren Spring Laboratory of DSIR.

Mr **T. F. Olt** has been elected Vice-President —Research of Armco Steel Corporation.

Dr **N. H. Polakowski** has been promoted to the rank of full professor at the Illinois Institute of Technology.

Mr **R. E. Race** has joined Blaw Knox Ltd. as Chief Metallurgist.

Major-General **E. P. Readman,** C.B.E., T.D., Special Director of The English Steel Corporation Ltd. and Managing Director of The English Steel Tool Corporation Ltd., has retired after 48 years' service with the Company. He has joined the Boards of Darwins Group Ltd. and Andrews Toledo Ltd., The Sheffield Forge and Rolling Mills Co. Ltd., and the other subsidiary companies of the group.

Mr **H. W. Riddervold** is now Managing Director of Forenede Nagle og Skruefabriker, Oslo.

Professor **G. Sachs** has been awarded the honorary degree of Doktor-Ingenieur by the Bergakademie Clausthal in recognition of his distinguished services to general and applied metallurgy. The Technische Universität, Berlin, has elected Professor Sachs Ehrensenator.

Mr **T. J. Smith** has been appointed a director of Colvilles Ltd.

Mr **G. Swinyard** is now Chief Metallurgist of The Phosphor Bronze Co. Ltd.

### Obituary

Mr **Joseph Elwell** (*elected* 1940), of Wolverhampton, on 18th April, 1958.

Mr **Luther Griffiths** (*elected* 1936), of Port Talbot, on 4th November, 1958.

Dr **James T. Mackenzie** (*elected* 1923), of Birmingham, Ala., U.S.A., on 17th November, 1958.

Mr **Frank Rutland Stagg** (*elected* 1938), of Sheffield, on 14th November, 1958.

Mr. **Walter Albert Louis Wolf,** B.SC. (HONS.), (*elected* 1945), of Kettering, on 7th March, 1958.

Professor Dr.Ing. **Frederick Zimerman** (*elected* 1949), of Kosice, Czechoslovakia, on 6th November, 1958.

## AFFILIATED LOCAL SOCIETIES

### Leeds Metallurgical Society

The Officers and Committee of the Society for the 1958–1959 session are as follows:

*President*
Mr K. Hill

| *1st Vice-President* | *2nd Vice-President* |
|---|---|
| Mr W. L. Bolton | Dr E. Simister |
| *Treasurer* | *Secretary* |
| Mr A. Lenton | Dr P. Feltham |

*Committee*

| | |
|---|---|
| Mr C. V. Wilson | Mr R. Mercer |
| Dr W. R. Berry | Mr E. W. Dewhirst |
| Mr F. K. Neath | Mr K. Dobson |

Professor N. J. Petch

*Past-Presidents*

| | |
|---|---|
| Mr C. Breckon | Mr A. Lenton |

Mr P. A. Green

## INSTITUTE OF METALS

### High-temperature Oxidation

Dr C. E. Birchenall, Associate Professor of Chemistry at Princeton University, U.S.A., has accepted an invitation from the Metal Physics Committee of The Institute of Metals to lecture on "The Oxidation of Metals at High-Temperatures." The lecture will be given at 6.30 p.m. on Thursday, 15th January, 1959, at the Institute's headquarters, 17 Belgrave Square, London, S.W.1. Visitors will be welcome; no tickets are required.

## BRITISH IRON AND STEEL RESEARCH ASSOCIATION

### 12th Chemists' Conference

The twelfth BISRA Chemists' Conference was held at Harrogate on 15th and 16th October, 1958, under the chairmanship of Dr E. Gregory (Edgar Allen and Co. Ltd.); Mr B. Bagshawe (Brown-Firth Research Laboratories) acted as deputy chairman. The attendance of 195 was a record.

The conference opened with a paper on the determination of toxic substances in the atmosphere, presented by Dr J. C. Gage (Industrial Hygiene Laboratories, I.C.I. Ltd.). Dr Gage reviewed the problems of setting limits for contamination in industrial atmosphere and described methods of collecting and analysing samples, referring in particular to instrumental and direct-reading methods that do not require the services of a skilled analyst.

Dr J. D. Hobson (Hadfields Ltd.) next spoke on the determination of gases in steel. After describing recent developments in methods and apparatus for determining hydrogen, nitrogen, and oxygen in steel, he went on to give examples of problems in which laboratory analysis had helped to solve works problems.

The second session opened with a paper on the determination of trace amounts of cobalt in iron and steel, presented by Mr J. L. West (Hadfields Ltd.). The method is of particular value in view of the Atomic Energy Authority's specifications.

A group of three papers on spectrographic methods for carbon, sulphur, and phosphorus followed. The first was presented by Mr J. H. Oldfield (Admiralty Materials Laboratory), who spoke on the vacuum ultra-violet region, which extends from approximately 1900 to 10Å. Mr Oldfield discussed the features of this region and its potentialities for the rapid determination of carbon, sulphur, and phosphorus.

Mr C. E. A. Shanahan (Richard Thomas and Baldwins Ltd.) presented a paper on the determination of carbon, sulphur, and phosphorus in plain carbon steels with the Hilger Fluorite Polychromator. He gave details of experiments on the influence of operating conditions on the sensitivity of the instrument and the reproducibility of the results. Experiences with the Hilger Vacuum Polychromator for routine control analysis were described in a paper by Mr P. Gale (Steel Company of Wales Ltd.), who compared the relative advantages and disadvantages of this three-element instrument and the various multi-channel direct-reading instruments that are now becoming available.

Methods for the determination of phosphorus in iron and steel were reviewed by Mr L. E. Gardner (Edgar Allen and Co. Ltd.) in the last paper in this session. Mr Gardner referred in particular to the work carried out during the last ten years by the Pig-iron, Carbon and Low-alloy Steels Analysis Subcommittee and the Physico-chemical Methods of Analysis Sub-committee of BISRA.

As is now customary, a paper of wider interest was presented in the third session. This year, Dr D. R. Harries (Atomic Energy Research Establishment) spoke on steels for nuclear applications. He first defined the basic factors that govern the choice of

materials for use in nuclear reactors and discussed the properties that are desirable in steels for the various components of a reactor. Dr Harries then gave an account of the effects of radiation on the properties of steels and discussed the types of steel that are likely to be required in nuclear plant of the future.

## 26th Blast Furnace Conference

The 26th Blast Furnace Conference was held at Harrogate on 8th and 9th October, 1958. The chair was taken by Dr T. P. Colclough (British Iron and Steel Federation), and Mr E. M. Summers (The Stanton Ironworks Co. Ltd.) acted as vice-chairman. The theme of the conference was a consideration of the factors that constitute blast-furnace input and their effect on output.

The first session was introduced by a paper from the chairman, in which he reviewed the effect of modern trends in operation of the blast-furnace. Dr Colclough described the effect on the heat balance of the blast-furnace resulting from the use of modern techniques such as higher blast temperatures, addition of steam to the blast, and oxygen enrichment.

Three papers, dealing with both the theoretical and the practical aspects of coke, were then presented. Mr J. P. Graham (British Coke Research Association) described recent research work on coke, Mr S. D. Carson (Dorman Long (Steel) Ltd.) described large-scale practical experimentation on coke quality, and Mr W. Hunter (Workington Iron and Steel Co. Ltd.) reviewed the effect of improved coke quality on furnace performance.

The second session was opened by papers from Mr J. S. McMahan and Mr J. W. Thomas (The Steel Company of Canada Ltd.) and Mr R. W. Johnson (South Durham Steel and Iron Co. Ltd.). Mr. McMahan reviewed the general effect of self-fluxing sinter on the working of the blast-furnace, and then went on to describe a 5-month trial using self-fluxing sinter on an 18-ft furnace at the Steel Company of Canada's works. Mr Johnson gave details of operating experience with the 22½-ft furnace at West Hartlepool following improvements made in their ore-preparation and sintering plant. In the ensuing discussion several speakers stated a preference for small-sized sinter; at one British works sinter was pre-screened to $-\frac{5}{16}$ in. before charging to the furnace.

The first part of the third session was devoted to a review of some of the external operating factors that could exert a beneficial influence on furnace performance. Mr E. W. Voice and Dr A. Grieve (BISRA) presented papers dealing with various developments and trends in blast composition. Both speakers gave very full descriptions of such items as injection of fuel at the tuyeres and higher blast temperatures, referring in particular to work on similar aspects abroad.

Finally two papers dealing with works experience on varying burden compositions were presented by Mr J. J. Evans (John Summers and Sons, Ltd.) and Mr W. Banks (Colvilles Ltd.). Each speaker described the make-up of the burden at his own plant, and the subsequent effect on furnace performance. Mr J. W. Houghton (Park Gate Iron and Steel Co. Ltd.) gave a paper dealing with the effect of slag composition on iron quality and output.

## BRITISH NUCLEAR ENERGY CONFERENCE

### Symposium on Geneva Conference

The draft programme of the Lecture Symposium on the Geneva Conference, to be held at Central Hall, Westminster on 13th March, is as follows:

**Morning Session**

(*Chairman:* Sir Josiah Eccles)

9.45–10.30: "Geneva 1958," by Sir John Cockcroft.

10.35–11.20: "Developments in physics," by Dr J. V. Dunworth.

11.25–12.10: "Power reactors," by Professor J. Diamond in collaboration with G. K. W. Ford, R. P. Kinsey, and R. H. Burdett.

12.15–1.00: "Developments in overall reactor technology," by Dr H. Kronberger.

**Afternoon Session**

(*Chairman:* Sir George Thomson)

2.30–3.15: "Metallurgical processes," by H. M. Finniston in collaboration with L. Grainger.

3.20–4.05: "Chemical processes," by Professor J. M. Kay in collaboration with A. S. White, C. Turner, K. Saddington, and Dr C. G. James.

4.10–4.55: "Thermonuclear processes," by D. W. Fry.

All the lectures will be published in full in the April 1959 *Journal of the B.N.E.C.* Admission is free to all members of the constituent societies without tickets.

## CONTRIBUTORS TO THE JOURNAL

**A. I. Aitken,** B.SC.—Works Manager, Dalzell Works, Colvilles Ltd. (*author of paper in December*, 1958, *issue*).

Andrew Aitken was educated at Queen's Park School and Glasgow University, graduating B.Sc. in metallurgy in 1940. He joined Colvilles Ltd. in the research department at Dalzell Works, and was engaged on production research in melting shops for two years. He then went to Mavisbank as manager of Colvilles' part of the works, returning to Dalzell three years later to take charge of furnace design and operation, fuel, and refractories. Since then he has been successively Deputy Chief (Technical and Quality Control), Production Manager, and, since 1957, Works Manager at Dalzell Works.

Mr Aitken has a distinguished sporting record. He has played football for Queen's Park, Third Lanark, and Hamilton Academicals, and represented Great Britain in the 1948 Olympic Games.

**P. R. V. Evans,** B.SC., PH.D.—Research Department, Metropolitan-Vickers Electrical Co. Ltd.

Peter Rees Vaughan Evans was born in 1932 and educated at St. Michael's School, Llanelly, and University College, Swansea. After graduating in 1953 he carried out research work under Professor O'Neill, for which he was awarded the degree of Ph.D. in 1947. In that year he joined Metropolitan-Vickers Research Department, where he is at present engaged on research in the field of nuclear power metallurgy.

**G. Hoyle,** B.SC. (HONS.), L.I.M.—Metallurgy (General) Division, British Iron and Steel Research Association, Sheffield.

Mr Hoyle was educated at Hipperholme Grammar School, Halifax, and at Leeds University, where he obtained a B.Sc. (Hons.) degree in metallurgy in 1947. He joined BISRA after graduating, and in recent years has specialized in problems connected with tool steels, and in particular with their structure and properties.

G. Hoyle

**P. L. Lillywhite,** B.SC.—Mechanical Engineering Section, Plant Engineering and Energy Division, British Iron and Steel Research Association (*author of paper in December*, 1958, *issue*).

Mr Lillywhite was educated at Wanstead County High School. He began studying for a London University degree in mechanical engineering while working in the laboratories of the Engineering Department of the South-West Essex Technical College. After working for a year in the design office of Keith Blackman Ltd., he joined the Royal Engineers for National Service; on being demobilized in 1954, he joined BISRA, and completed his degree the following year. Mr. Lillywhite has recently been engaged in investigating forging and rolling plant problems.

**J. H. Stoll**—Chief Metallurgist, Bethlehem Steel Company.

John Howard Stoll is a native of Camden, New Jersey, and graduated as a metallurgical engineer from Lehigh University. He has been associated with the Bethlehem Steel Company for over 36 years, and was appointed Chief Metallurgist in 1945. As such, he is responsible for all phases of quality control at the Bethlehem plant, covering such factors as specifications, observations, laboratories, metallurgical contact, and product inspection, and to this end his wide practical experience in all the many branches of the plant, which produces structural steel, alloy and tool steel bars, and a variety of foundry, drop forge, press forge, and machine-shop products, is of great value.

Mr Stoll is a member of the American Iron and Steel Institute, the American Society for Metals, the American Society for Testing Materials, the American Ordnance Association, and the United States Naval Institute. He holds the United States Navy Bureau of Ordnance Award.

## EDUCATION

### Mond Nickel Fellowships

The Mond Nickel Fellowships Committee announced recently the award of Fellowships for 1958 to the following:

G. H. LONGWORTH (Lancashire Steel Manufacturing Co. Ltd.) to study the manufacture of plain carbon and alloy steels for

A. I. Aitken

P. R. V. Evans

P. L. Lillywhite

rod and wire products, in the UK, on the Continent, and in the USA.

J. C. MORRISON (Carron Company) to study melting, moulding, and heat-treatment of alloy cast irons, quality control of metal and sand plants in modern foundries, and new developments in related metallurgical processes.

T. TAIT (Colvilles Ltd.) to study the technical and economic factors involved in the use of oxygen in steel industries in the UK, on the Continent, and in the USA and Canada.

The Mond Nickel Fellowships Committee now invites applications for Fellowships of an approximate value of £900 to £1200 for 1959. Fellowships will be awarded to selected candidates of British nationality with degree or equivalent qualifications to enable them to obtain wider experience and additional training in industrial establishments, at home or abroad, to make them more suitable for future employment in senior technical and administrative positions in British metallurgical industries. Each Fellowship will cover one full working year. Applicants will be required to state details of the programme they wish to carry out. Particulars and forms of application are available from The Secretary, Mond Nickel Fellowships Committee, 4, Grosvenor Gardens, London, S.W.1. Completed application forms are required by 1st June, 1959.

## NEWS OF SCIENCE AND INDUSTRY

### Record Mill " Changeover " at Abbey

Flags were flown at the Abbey Works of The Steel Company of Wales Ltd. at Margam on 29th November as a tribute to 1 000 men who beat, by 1½ days, the 14-day target for installing a new rolling mill. The men, working in three shifts round the clock, made possible the start of production rolling on a new £1¾ million universal mill that will eventually increase the slab rolling output of the works to 60 000 tons per week.

When a £50 million development scheme, at present under construction, is completed about the middle of next year, steel ingot production will increase from 48 000 to 60 000 tons per week. The universal mill is needed to roll this extra output, for not only will it roll the ingots faster than the old mill, but it will also produce a better-quality slab.

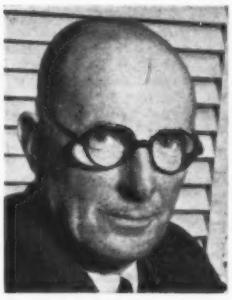
As all the ingots produced pass through the slabbing mill, the decision to make this changeover made it inevitable that there would be some interference with production while the work was in progress. For this reason the job had to be planned for completion in the shortest possible time: 14 days was allocated for it.

Success has been achieved thanks to most careful planning over three years, covering every item of the preliminary preparations and the changeover operation itself. All personnel were briefed in the minutest detail with the aid of models of both the old and the new mills, which could be dismantled and erected to demonstrate what each man was required to do. Details of the planning were given in a paper by C. E. H. Morris and R. N. Dale in a paper in the *Journal* (April, 1957).

Dismantling of the old mill started on 17th November, and was completed within two days. During this time 1 374 tons of equipment were removed. Installing the universal mill involved the erection of 2 375 tons of equipment, some items of which weighed as much as 142 tons.

Most of the electrical work was completed in advance of the changeover, and so the final electrical installation and the testing and commissioning of the mill were reduced to the shortest possible time.

Although it was necessary to reduce the iron and steel output of the works during this period, the hot strip mill was able to continue rolling, using slabs purchased in advance from other British steelmakers.



G. P. Tinker

J. C. Howard

F. S. Leigh

### Birlec–Efco Board

It has been announced that the Board of the newly formed company, Birlec–Efco (Melting) Ltd. (*J.I.S.I.*, 1958, Nov., p. 315) will consist of G. P. Tinker (*Chairman*), J. C. Howard (*Managing Director*), F. S. Leigh (*Assistant Managing Director*), D. L. Campbell, J. H. Crossley, P. F. Hancock, J. A. Monks, and T. G. Tanner. The Secretary will be J. C. Mantell.

The new Company's offices are at Westgate, Aldridge, Staffs. (telephone Aldridge 52071).

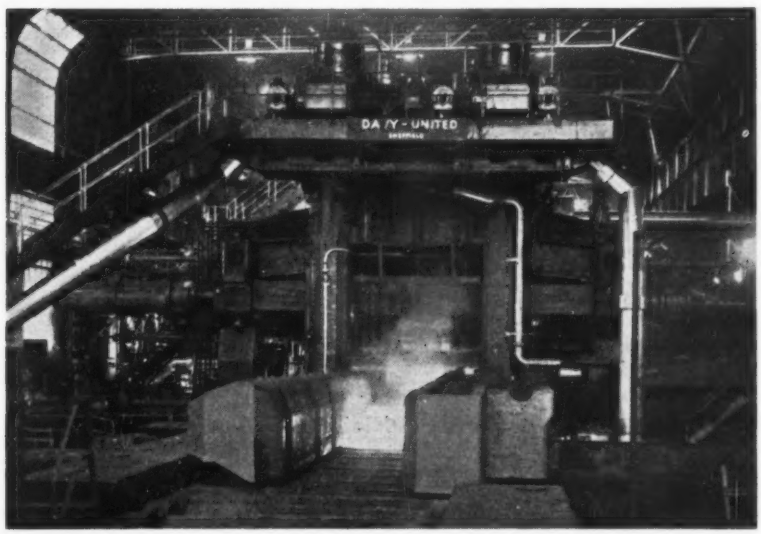
### PEP and the Steel Industry

PEP (Political and Economic Planning) has devoted Planning Broadsheet No. 428 to "The Development of the Steel Industry." The chapter headings—The Background, How much Steel?, Finance and Price Policy, Problems of Location, and Freer Trade in Steel—give some idea of the scope of this interesting 16-page booklet. It brings together the basic economic issues facing the steel industry, "no matter who owns it and whatever form of supervision is employed." These are the advance planning of steel capacity, the relation of pricing policies and decisions about location to the planning of development, and the impact on the industry of a possible European Free Trade Area.

Copies may be obtained from PEP, 16 Queen Anne's Gate, London S.W.1, price 2*s.* 6*d.*

### Corrosion Exhibition

The Corrosion Group of the Society of Chemical Industry is holding an exhibition relating to the prevention of metallic corrosion at the Battersea College of Technology, Battersea Park Road, London, S.W.11, on Thursday and Friday, 22nd and 23rd January. The main theme of the exhibition is the apparatus and techniques used for research and for the control of anti-corrosion processes. Attendance on the Thursday is restricted to members and their friends attending the Society's annual conversazione, and tickets will be required, but on the Friday it is open to all, without tickets, from 9.30 a.m. to 3.30 p.m.




## DIARY

**5th Jan.**—CLEVELAND INSTITUTION OF ENGINEERS—"*Blast-furnace techniques*," by Dr T. P. Colclough—Cleveland Scientific and Technical Institution, Middlesbrough, 6.30 p.m.

**6th Jan.**—INSTITUTION OF ENGINEERING INSPECTION—"*An introduction to engineering metallurgy*," by C. C. Griffin—Coventry Technical College, 7.30 p.m.

**6th Jan.**—SHEFFIELD METALLURGICAL ASSOCIATION (Methods of Analysis Group)—Annual General Meeting—BISRA, Hoyle Street, Sheffield, 7 p.m.

**6th Jan.**—SOCIETY OF CHEMICAL INDUSTRY (Corrosion Group and Birmingham and Midland Section)—"*Alloys for high-temperature service*," by D. A. Oliver, C.B.E.—Birmingham and Midland Institute, Paradise Street, Birmingham, 6.30 p.m.

**7th Jan.**—NEWPORT AND DISTRICT METALLURGICAL SOCIETY—"*The historical development of the rolling mill*," by E. C. Larke—Whitehead Institute, Newport, 7 p.m.

**8th–9th Jan.**—British Society of Rheology—Conference on Fracture, Fatigue, and Failure—University of Leeds.

**8th Jan.**—Liverpool Metallurgical Society (Joint Meeting with the University of Liverpool Metallurgical Society)—Student Prize Competition—Department of Metallurgy, University of Liverpool, 7 p.m.

**12th Jan.**—Lincolnshire Iron and Steel Institute—"*Failure of a 60-MW steam turbo-generator at Uskmouth power station*," by A. L. G. Lindley—North Lindsey Technical College, Scunthorpe, 7.30 p.m.

**13th–15th Jan.**—British Iron and Steel Research Association—Conference on Iron and Steelworks Buildings—Ashorne Hill, Leamington Spa.

**13th Jan.**—Institution of Engineering Inspection—"*Heat-treatment of steel*," by N. Tommis—Leeds Church Institute, 7.30 p.m.

**13th Jan.**—Sheffield Metallurgical Association (Refractories Group)—Annual General Meeting—BISRA, Hoyle Street, Sheffield, 7 p.m.

**14th Jan.**—Institution of Engineering Inspection—"*The application of powder metallurgy to the production of engineering components*," by Dr P. R. Marshall—Birmingham Exchange and Engineering Centre, 6.30 p.m.

**14th Jan.**—North Wales Metallurgical Society—"*Heat-resisting steels*," by W. E. Bardgett—Flintshire Technical College, Connah's Quay, 7 p.m.

**14th Jan.**—Swansea and District Metallurgical Society—"*Ingots and ingot moulds*," by G. N. Cherry and D. R. Fardoe—Central Library, Swansea, 7 p.m.

**15th Jan.**—British Nuclear Energy Conference—"*Electrical installations at Calder Hall Power Station*," by N. J. Mackay and E. Hardwick—Institution of Electrical Engineers, Savoy Place, London, W.C.2, 5.30 p.m.

**15th Jan.**—Ebbw Vale Metallurgical Society—"*Mental mechanisms*," by Dr W. Grey-Walter—R.T.B. Lecture Rooms, 7.15 p.m.

**15th Jan.**—Leeds Metallurgical Society—"*The ultimate structure*," by T. J. Parry and S. Newton—Chemistry Wing, University of Leeds, 7.15 p.m.

**15th Jan.**—Staffordshire Iron and Steel Institute—"*Continuous reheating furnaces*," by G. Harrison—Stoke-on-Trent, 7.30 p.m.

**16th Jan.**—Society of Chemical Industry (Corrosion Group and Glasgow Section)—"*Protection of motor vehicles from corrosion*," by Dr S. G. Clarke—Royal College of Science and Technology, Glasgow, 7.15 p.m.

**19th Jan.**—Sheffield Society of Engineers and Metallurgists—Annual General Meeting and Presidential Address—Sheffield University, 7.30 p.m.

**20th Jan.**—Sheffield Metallurgical Association—Annual General Meeting and Presidential Address—BISRA, Hoyle Street, Sheffield, 7 p.m.

**21st Jan.**—Manchester Metallurgical Society—"*The investigation of metallurgical failures in aircraft*," by D. A. Ryder—Central Library, Manchester, 6.30 p.m.

**22nd–23rd Jan.**—British Nuclear Energy Conference—Two-day Symposium on Nuclear Fuel Cycles.

**22nd–23rd Jan.**—Society of Chemical Industry (Corrosion Group)—Corrosion Exhibition—Battersea College of Technology, Battersea Park Road, London, S.W.11.

**27th Jan.**—North-East Metallurgical Society—"*Some characteristics of high-strength aluminium alloys*," by G. Goddard—Cleveland Scientific and Technical Institution, Middlesbrough, 7.15 p.m.

**27th Jan.**—Sheffield Metallurgical Association (Methods of Analysis Group)—"*The chlorination method for the separation of inclusions*," by F. J. Armson—BISRA, Hoyle Street, Sheffield, 7 p.m.

**28th Jan.**—British Nuclear Energy Conference—"*The design of electromechanical auxiliaries directly associated with power-producing reactors*," by A. E. Harwood, P. Scott, and B. H. Stonehouse—Institution of Electrical Engineers, Savoy Place, London, W.C.2, 5.30 p.m.

**30th Jan.**—Manchester Association of Engineers—"*Powder metallurgy*," by Dr L. Harrison—Engineers Club, Albert Square, Manchester, 6.45 p.m.

**2nd Feb.**—Cleveland Institution of Engineers—"*The design and construction of steam generators for nuclear power stations*," by F. Carr—Cleveland Scientific and Technical Institution, Middlesbrough, 6.30 p.m.

**3rd Feb.**—Sheffield Metallurgical Association—"*The production of heavy castings*," by J. H. Pearce—BISRA, Hoyle Street, Sheffield, 7 p.m.

**4th Feb.**—Manchester Metallurgical Society—"*Micro-mechanism of fracture*," by Professor W. S. Owen—Central Library, Manchester, 6.30 p.m.

**4th–7th Feb.**—National Metallurgical Laboratory—Symposium on the Iron and Steel Industry in India—Jamshedpur, India.

**6th Feb.**—Staffordshire Iron and Steel Institute, Annual Dinner, Wulfrun Hall, Wolverhampton.

**7th Feb.**—Swansea and District Metallurgical Society—"*Recent developments in cast iron*," by I. C. H. Hughes—Central Library, Swansea, 6.30 p.m.

## BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on p. 436 of the December, 1958, issue of the Journal:

917. Altykis, A. V., and N. I. Belan: "Raising the Temperature of Final Deformation as a means of reducing the Force required for Forging Large-size Billets." *Metallov. Obra. Met.*, 1957, Oct., pp. 48–52. (£2 15*s*. 0*d*.)

968. Fomichev, I. A.: "Stress Conditions in Oblique Rolling." *Stal'*—Rolling and Tube Manufacture Supplement, 1958, pp. 176–206. (£7)

982. Shimanaka, K., *et al.*: "On the Desulphurisation of Molten Steel by $CaC_2$ and CaO." *Tetsu-to-Hagane*, 1956, Sept., pp. 864–865. (£1)

1000–1008. *Papers presented at the International Transport Conference, June 1958, Liège.*

1000–1002: Cost.

1000. Forterre, R.: "The Main Element of the Cost of Internal Transport: Charging." (£5)

1001. Mévis, N.: "Operating Costs of the Internal Transport Services of Iron and Steel Works." (£3 10*s*. 0*d*.)

1002. Mieth, H.: "A Comparison of the Performance, Lay-out and Cost of Different Works Railway Systems of German Foundries supplemented by Mathematical Statistics." (£5)

1003–1005. Maintenance.

1003. Willem, P.: "Steelworks Railway Maintenance: New Methods and Mechanisation." (£8 5*s*. 0*d*.)

1004. Forterre, R.: "Technical and Economic Aspects of Track Maintenance in a Steelworks." (£3 15*s*. 0*d*.)

1005. Müller, A.: "Modern Methods and Procedures for the Maintenance of Works Railway Track." (£6 10*s*. 0*d*.)

1006–1008: Handling.

1006. Gesell, W.: "Transport of Lime and Scrap in an Open-hearth Furnace Steelworks." (£6)

1007. Lafon, A. F. de, and G. Gabrielle: "Handling of Lime and Scrap in Steelworks." (£4)

1008. Hubin, L.: "Handling of Lime and Iron Scrap at Cockerill-Ougrée." (£2 15*s*. 0*d*.)

1040. Wolfstieg, U.: "On the Accuracy of Stress Measurements by X-ray Methods, particularly in the Case of Wide Lines." *Acta Crystallographica*, 1958, March, pp. 45–48. (£2 10*s*. 0*d*.)

1051. Wilz, G., and M. Petitdidier: "Continuous Casting and its Problems." *Rev. Mét.*, 1958, (5), pp. 486–494. (£4 5*s*. 0*d*.)

1053. Möbius, H. E., and F. Pawlek: "Cube Orientation as Recrystallization Texture in Iron–Silicon Alloys." *Arch. Eisenhüttenwesen*, 1958, July, pp. 423–432. (£6)

1055. Wolf, W., *et al.*: "Experiments carried out in the Plant on the Smelting of Conakry Ore." I—'Smelting Conakry Ore in the Blast Furnace.' *Stahl u. Eisen*, 1958, July 24th, pp. 1020–1027. (£5 10*s*. 0*d*.)

1056. Wolf, W., *et al.*: "Experiments carried out in the Plant on the Smelting of Conakry Ore." II—'Extraction of Chromium from Pig-iron by refining with Pure Oxygen in the Ladle.' *Stahl u. Eisen*, 1958, Aug. 7th, pp. 1100–1107. (£5 10*s*. 0*d*.)

1068. Becker, G.: "Corrosion Resistant Steel Bolts." *Draht*, 1958, Jan., pp. 13–15. (£2 5*s*. 0*d*.)

1070. Koch, W., and A. Schrader: "Change in the Structure of Austenitic Steels in Creep Rupture Tests." *Stahl u. Eisen*, 1958, Sept. 4th, pp. 1251–1262. (£6 10*s*. 0*d*.)

1076. Maksimovich, G. S.: "Improved Methods for the Manufacture of Hard Alloy Tube Drawing Dies." *Stal'*, 1958, July, pp. 629–633. (£4 5*s*. 0*d*.)

1077. Teterin, P. K.: "An Analysis of Existing Methods of Roll Profiling for Hot and Cold Pilger Mills." *Stal'*,—Rolling and Tube Manufacture Supplement, 1958, pp. 227–242. (£4 10*s*. 0*d*.)

1081. Levelink, H. G.: "Some Characteristics of Waterglass as a Binder for Moulding and Core Sands." *Metalen*, 1957, Oct. 31st, pp. 406–412; Nov. 15th, pp. 442–444. (£5)

1083. Ryzhkov, P. Ya., *et al.*: "Properties of Oxygen-blown Converter Steel." *Stal'*, 1958, July, pp. 643–647. (£3)

1085. Teterin, P. K.: "An Analysis of Friction Forces and Roll Engagement in Disc Mills." *Stal'*,—Rolling and Tube Manufacture Supplement, 1958, pp. pp. 207–226. (£5)

1087. Glinkov, G. M.: "Control of Thermal Conditions by maintaining Maximum Heat Absorption Capacity of the Open-hearth Bath." *Stal'*, 1958, Apr., pp. 370–376. (£5 5*s*. 0*d*.)

1089. Niklas, L.: "A New Method for Thickness Measurement (Ultrasonic Velocity Measurement)." *Z. Metallkunde*, 1958, March, pp. 152–155. (£2 10*s*. 0*d*.)

1092. Hendus, H., *et al.*: "Particle Size and Grating Distortion of Tungsten Powders and Tungsten Carbide Powders." *Z. Metallk.*, 1957, Dec., pp. 615–624. (£5 5*s*. 0*d*.)

1096. Latour, A., and L. Heinen: "Practical Results in the Basic Converter with the Use of Small Sized Lime made in the Cross-flow Shaft Kiln." *Stahl u. Eisen*, 1957, Apr. 4th, pp. 426–428. (£2 10*s*. 0*d*.)

1114. Menuet-Guilbaud, B., *et al.*: "Influence of Suction on the Sintering Process." (35 pp.) 2nd International Symposium on the Sintering of Iron Ores, 1957. (£6 5*s*. 0*d*.)

1133. Möltgen, G.: "The Reactive Power of Three-phase Current Converters with Unsymmetrical Control." *Arch. Eisenhüttenwesen*, 1956, May, pp. 272–285. (£3 5*s*. 0*d*.)

1134. Duflot, J., and S. Dérot: "Investigation on Slab Moulds of More than Three Tons." *Fonderie*, 1958, Oct., pp. 467–474. (£10)

1135. Demaret, M.: "Special Blowing Techniques [using $O_2$–$CO_2$] at the Basic Bessemer Shop of the 'Aciéries et Minières de le Sambre' Works at Monceau-sur-Sambre." Journées Internationales de la Sidérurgie, 1958, June, pp. 1–9. (£3 15*s*. 0*d*.)

# *ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES*

## Iron and Steel Manufacture and Related Subjects

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## ABBREVIATIONS AND LIST OF PERIODICALS EXAMINED BY THE ABSTRACTING SERVICE

### Additions and alterations to the January, 1958, list

*Delete from the list the following periodicals, which are discontinued or have changed their names:*

Atomics and Nuclear Energy (London)
Bulletin of the Institute of Metal Finishing (London)
Chemical Engineering News (Washington)
Demag News (Duisburg)
Elliott Journal (London)
Iva Ingeniörs Vetenskaps Akademien (Stockholm)
Industrial Heating Engineer (London)
Insulation Review (London)
Journal of the Institution of Heating and Ventilating Engineers (London)
Mineralogical Magazine (London)
Prace Instytutow Minsterstwa Hutnictwa (Stalinogrod)
Steel Processing (Pittsburgh)
Techniques et Civilisations (St. Germain-en-Laye)
Western Metals (Los Angeles)

*Insert into the list:*

**Abstracts** Journal **Met**allurgy (New York, London)
**Acta Chim**ica **Sin**ica (Peking)
**Acta Met**allurgica **Sin**ica (Peking)
**Acta Polytech**nica **Scand**inavica (Copenhagen, Helsinki, Oslo, Stockholm)
**Acta Technica** (Prague)
**Anglo-Swedish Rev**iew (London)
**Atomic World** (London)
**Battelle Tech**nical **Rev**iew (Columbus, Ohio)
**Bibliografia** della **Saldatura** (Rome)
**Bull**etin de la **Soci**été **Chim**ique **Beograd** (Beograd)
**Centre D'inf**ormation du **Chrome Dur,** Bulletin de Documentation (Paris)
**Chem**ical **Ind**ustry & **Eng**ineering (Peking)
**Chinese** Journal of **Mech**anical **Eng**ineering (Peking)
**Colorado School** of **Mines**, Mineral Industries **Bull**etin (Colorado)
**Demag Nachr**ichten (Duisburg)
**Dodatok** do **Ukrain**'skogo **Fiz**ichogo **Zhur**nalu (Kiev)
**Economist** (London)
**Ergonomics** (London)
**Heating** (London)
**Huttenzeitung** (Bochum)
**Inco-Mond Mag**azine (London)
**Iron** and **Steel Rev**iew (Calcutta)
**Ironworker** (Lynchburg)
Journal of Labour **Hygiene** in **Iron** and **Steel Ind**ustry (Tokyo)
**Kupfer Mitt**eilungen (Berlin)
**Met**al **Phys**ics (Tokyo)
**Mineria** y **Met**alurgia (Madrid)
**Mitt**eilungen der **Forschungsges**ellschaft **Blechverarb**eitung e.v. (Dusseldorf)
**Monthly Rep**ort of the **Iron** and **Steel Fed**eration (Tokyo)
**New Zea**land Journal of **Science** (Wellington)
**Nuclear Sc**ience **Abs**tracts (Oak Ridge, Tennessee)
**Official J**ournal (Patents) (London)
**Osterr**eichische **Ing**enieur **Zeit**schrift (Vienna)
**Periodica Polytech**nica (Budapest)
**Physical Rev**iew **Letters** (New York)
**Physics** of **Met**als and **Metallog**raphy (New York, London)
**Powder Met**allurgy (London)
**Powder Met**allurgy **Quart**erly (New York)
**Prace** Brněnské **Zakladny** Ceskoslovenské Akademie věd (Prague)
**Prace Instytutow Hutniczych** (Katowice)
**Rev**iew of the **Polish Acad**emy of **Sci**ences (Warsaw)
**Revue M Tijds**chrift (Brussels)
**Rhodesian Min**ing & **Eng**ineering (Salisbury)
**Rivista Ital**iana Della **Saldatura** (Genoa)
**Rudy** i **Met**ale **Niezelazne** (Katowice)
**Safety** (London)
**Scientia Sin**ica (Peking)
**Solus Schall News** (London)
**Technique Mod**erne (Paris)
**TVK** Teknisk-Vetenskaplig Forskning (Stockholm)
**Times Lit**erary **Supp**lement (London)
**Toyó Kôhan** (Tokyo)
**Trans**actions of the **Met**allurgical Society of **AIME** (New York)
**Transport Age** (London)
**Ukrains'kii Fizichnii Zhur**nal (Kiev)
**Unesco Bull**etin for **Libraries** (Paris)
United States Government **Res**earch **Rep**orts (Washington)
**Werkstatt** und **Masch**inenbau (Berlin)
**Western Metalw**orking (Los Angeles)
**Zhur**nal **Eksper**imental'noi i **Teoret**icheskol **Fiz**iki (Moscow)

## ABSTRACTS

### MINERAL RESOURCES

**The Extrusive Sedimentary Ore Deposits of the Eastern Alps. Part I—Magnesite and Hematite Deposits.** F. Hegemann. (*Z. Erz. Met.*, 1958, **11**, May, 209–217). The geochemistry of magnetites and hematites is outlined, and is followed by a discussion of the primary origins of the magnetite and hematite deposits of the Eastern Alps, with reference to a number of examples, and taking into account views expressed in the literature. (29 refs.).—L. D. H.

**Iron Ore in the Area of Upper Silesian Industrial District.** T. Szreter. (*Hutnik*, 1957,

**24,** October, 397–401). [In Polish]. The distribution of iron ore in the area of Upper Silesia is surveyed. Blackband ironstone and its mining in Germany and England are discussed. The need for further research into the assessment of ferrous carbonate ores is emphasized.

**The Iron Ore Supply Situation in Europe.** H. Putzer. (*Eng. Min. e. Met.*, 1958, **27,** Feb., 73–78). [In Portuguese]. The author describes present production and imports of the European iron and steel producing countries. Future developments and trends are discussed with particular reference to the ore exporting countries. Brazil could export up to 30 million tons of high grade ore yearly once her production and transport difficulties are solved.

**Geology of the Manganese Deposits of Cuba.** F. S. Simons and J. A. Straczek. (*U.S. Geol. Survey, Bull.*, 1958, (1057), pp. 289).

**Manganese and Chrome Ore Outlook.** J. M. Warde and Eileen P. Burke. (*Min. Eng.*, 1958, **10,** Aug., 868–873). Estimates of demand and supply, including the U.S.S.R. are given.

## ORES—MINING AND TREATMENT

**Iron Ore.** H. T. Reno. (*U.S. Bur. Mines Min. Yearbook* 1955, **1,** 547–577). Statistics of mining, transportation and use are given.

**The Arctic Iron Ore District of Sweden.** M. Winkler. (*Neue Hütte*, 1958, **3,** June, 331–340). A brief description of the main ore deposits of Sweden is followed by a detailed account of the mines and ore preparation plants of the Arctic district.

**Contribution to the Choice of Techniques for Preparing Lorraine Ores.** Ch.-G. Thibaut. (*Rev. Mét.*, 1958, **55,** June, 513–523). From an analysis of the performance of a large number of blast furnaces, a method is developed for calculating burden and other factors. Data are calculated for various ways of preparing the burden using self-fluxing Lorraine ores. A numerical example is given.

**Attempts to Enrich Certain Manganiferous Lean Iron Ores.** D. Totoescu and A. Steclaci. (*Studii si Cercetari de Metalurgie*, 1958, **3,** (1), 53–67). [In Rumanian]. Laboratory scale trials have been carried out with the object of establishing conditions for extracting iron from the deposits in the Vascău-Moneasa region. These ores are low in iron, and high in silica, alumina and manganese. It is considered that the ore could be treated by the Krupp–Renn process, with the addition of dolomite.

**New Canadian Processes for Low-Grade Iron Ores.** (*Min. J.*, 1958, **251,** Sept. 12, 280–282). Work by the Ontario Research Foundation is described. Dry grinding and concentration, flash-roasting, jet-smelting of dry concentrates and a pelleting machine are outlined.

**New Process, New Plant: High Grade Iron from Inco's Concentrates.** (*Min. Eng.*, 1958, **10,** Aug., 864–866). An account of the treatment of nickel ore residues by roasting off the S, reducing in kilns in co-current gas-solid flow, leaching out the Ni with $NH_3$ solution, magnetic thickening and agglomeration of fines, which are fired on a travelling grate.

**J & L Installs Sintering Plant at Otis Works.** (*Iron Steel Eng.*, 1958, **35,** March, 174–176). A new sintering plant with a daily capacity of 2500 t is described. The plant cost $7m. It has a Dwight-Lloyd travelling bed of material 8 ft. wide and 128 ft long. The material may be 12 to 16 in. deep.—M. D. J. B.

**The Influence of Iron Coke on Refractory Materials.** M. Starczewski. (*Prace Inst. Hutn.*, 1958, (2), 77–89). General practical and theoretical problems connected with the corrosion of ceramic materials associated with iron coke are considered and an investigation is made of materials to determine their corrosion resistance during the formation of iron coke of varying iron contents at temperatures from 900–1260° C. The resistance of chrome-magnesite to the action of iron coke is for all practical purposes complete over the whole range of temperatures in which experiments were carried out. The few rust strains observed on the surface of the specimens were caused by weak infiltration of iron into the bottom of the pores and did not cause the slightest change in structure.

Both types of siliceous materials resisted the action of iron coke with a content of up to 50% Norwegian magnetite ore at up to 1170°. The action of iron coke consisting of 75% Norwegian ore, blast furnace dust or burnt pyrites reduced the limit of refractoriness to 1140°. Quartzite-chamotte materials are completely resistant to this action up to 1140° (content 75% Norwegian ore, blast furnace dust or burnt pyrites), but their use for lining coke batteries is precluded by strong infiltration of carbon monoxide and resultant structural changes. Chamotte materials did not show corrosion as a result of the coking process of mixtures of coal with up to 75% magnetite, burnt pyrites or blast-furnace dust up to 1140°. At this temperature, however, changes are already visible, disclosed by the grey discoloration of the products, which may show the occurrence of infiltration of CO, inclusion of C and occurrence of a Bell reaction. It must be emphasized that the temperatures attained are regarded as safe for the individual refractories, and the coking mixtures as not harmful so long as no structural changes are observed in the material. This proviso arises from the difficulty of foreseeing, on the basis of laboratory experiments, their behaviour in coke batteries where they will be subjected to corrosion over many years. Analysis of the results shows that apart from the corrosive action of iron coke, the quartzite-chamotte materials show a considerable increase of porosity and a reduction of bulk density after repeated contact with iron and coke. This is probably connected with the catalytic influence of iron oxides on oxidation (gas corrosion) under the influence of CO.—M. T.

**Investigation of Mineral Content of Sinter from Krivoirog Iron Ore.** E. F. Vegman. (*Zavods. Lab.*, 1958, (4), 444–446). A detailed analysis is here presented of the chemical content of these ores, especially regarding the difficult calculation of the magnetite and wustite proportions, and the results are tabulated.—L. H.

**Method for Determining the Softening Point of Ores and Sinter.** D. G. Khokhlov and Yu. A. Gyrdymov. (*Zavods. Lab.*, 1958, (4), 443–444). To avoid errors due to the interaction of materials with the walls of iron or porcelain containers, the present method proposes the use of briquettes of the materials prepared from 1–0 mm grain size samples compressed under a load of 150 kg/cm² to suitable dimensions. A description of the apparatus and of the method is given in detail.

## FUEL—PREPARATION, PROPERTIES AND USES

**Paradoxes of Lorraine Coals Used in Coking Plants.** J.-H. Grandval. (*Chal. Ind.*, 1958, **39,** June, 145–149). A lecture dealing with some unusual features associated with the use of low-rank Lorraine coals for coking.

**Physical and Chemical Changes Associated with the Transformation of Semi-Coke into Coke in the Temperature Range 500–1000° C.** A. Kijewska. (*Prace Inst. Hutn.*, 1958, (2), 124–132). An investigation was made of the carbonization of various grades of coal from 4 Polish coalfields in respect of content of voltailes and ash, apparent and true density, mechanical strength, shrinkage, pore wall thickness, degree of leanness obtained and degree of carbonization (Carbon saturation), and relationships were established between these factors and the coking temperature. A detailed analysis was made of the relationship between the real density and the strength of the actual coke substance. Pore wall thickness is reduced in proportion to the increase in temperature, but without apparent loss in strength. A relationship was also established between the degree of carbon saturation, the real carbon density and the voltaile content of the coke.—M. T.

**The New " Gilsonite " Coke for the Electro-metallurgical Industries.** American Gilsonite Co. (*J. Four Elect.*, 1957, **62,** (5), 167–171). An account is given of the production of motor fuel and high-grade coke from gilsonite, and of the industrial applications of the coke.

**The Treatment of Water for Use in Coke Oven and By-Product Plants.** F. Walsh. (*Year Book Coke Oven Man. Assoc.*, 1958, 203–215).

## AIR POLLUTION AND SMOKE

**Practical Hints on the Gravimetric Determination of the Dust Content of Flowing Gases, with Special Reference to the Cyclone Probe.** E. Walter. (*V.D.I.Z.*, 1958, **100,** June 11, 736). A brief outline of the use of the cyclone probe in determining the dust content of flowing gas, with a diagram of the apparatus.

**Model Studies Expect to Improve Steel Plant Precipitator Operation.** (*Iron Steel Eng.*, 1958, **35,** March, 181–186). Means of improving precipitators and preventing air pollution are discussed. Particular mention is made of O.H. sintering, hot scarfing cupola and oxygen conversion precipitators.

## TEMPERATURE MEASUREMENT AND CONTROL

**Thermocouples with Non-Linear Characteristics Composed of Fe–Al–Cr and Cu–Ni Alloys.** Y. Shirakawa, T. Ohara and D. Amemiya. (*Nippon Kinzoku*, 1956, **20,** Mar., 131–135). Thermal e.m.f.'s were measured for Fe–Al, Fe–Cr, Fe–Al–Cr and Cu–Ni couples between 100° and 800° C. A couple formed between Cu–1·94% Ni and Fe–6·2% Al–0·94% Cr alloys has an e.m.f. which is almost zero between room temp. and 400° C, and then increases to 5·4 mV at 800° C.—K. E. J.

**Investigations of the Development of an Oxide Thermocouple.** W. A. Fischer and G. Lorenz. (*Arch. Eisenh.*, 1958, **29,** May, 293–300). Investigations are described of the properties of the thermocouple Pt/$Cr_2O_3$/$Cr_2O_3$ + 1 mol % $TiO_2$/Pt, with the object of developing a thermocouple operating at high temperatures, and offering high resistance to chemical attack. Measurements of thermo-electric power and potential showed the couple to be suitable for high temperature measurements. Thermoelectric power and potential are substantially greater than those of metallic couples, and the former is very reproducible, remaining constant within permissible limits of error even after heating for 24 h at high temperatures. Thermoelectric potentials are given at intervals of 10° C up to 1750° C, and obtained by extrapolation up to 2100° C.

## REFRACTORY MATERIALS

**A Basic Insulating Refractory.** H. M. Harris and H. J. Kelly. (*Amer. Ceram. Soc. Bull.*, 1958, July, 307–311). Porosity is produced by the action of $H_3PO_4$ and ammonium ligninsulphonate on MgO. The product sets in 3 min and can be handled in 20. Porosities of 70%, bulk densities of 70 lb/ft³ and modulus of rupture of 350 psi. can be attained.

**The Vitreous Phase in Refractory Materials.** M. Kantzer. (*Silicates Indust.*, 1958, **23,** Apr., 185–190). The conditions of formation of a vitreous phase in refractory material with particular reference to alumino-silicate and refractory oxide materials are discussed. (16 references).—B. G. B.

**Bond Strength, Radioactive Tracers and Casting-Pit Refractories.** D. H. Houseman. (*Refract. J.*, 1958, **34,** Aug., 353–361, 364; discussion, 362–364). The hot compressive strength test is applied to a number of materials. In the second part, the location of the source of non-metallic inclusions in steel is studied using $^{140}$La in furnace slag and $^{140}$Ba in ladle bricks. No significant amounts of inclusions from either source were found.

**A Study of the Problem of Calcium Silicide Disintegration.** B. Paczula and Z. Kulinski. (*Prace Inst. Hutn.*, 1958, (2), 91–94). The authors examine the process of Ca–Si pro-

duction, the phenomenon of the breakdown of the material which sometimes occurs even during the production process, and investigate its causes. The main factor is shown to be the inclusion of slag impurities, and the mechanism of the action of the slag is examined, and methods are proposed for avoiding slag contamination. The influence of casting height on the stability of Ca–Si is investigated; the cascade casting method described eliminates slag impurities. Research was also carried out into the considerable scrap caused by the breakdown of Ca–Si in the production of electrical sheet steel, through the combination of $CaC_2$ with atmospheric moisture to form $C_2H_2 + CaO$, the gas hereby evolved causing cracking of the calcium silicide.—M. T.

**The Work of CECA on Silica Roofs.** J. Baron. (*Centre Doc. Sider. Circ.*, 1958, **15**, (5), 1071–1072). A short discussion of factors affecting the performance of silica refractory material is presented.—B. G. B.

**The Effect of the Addition of a Ferrous Mineraliser to Siliceous Refractory Products on their Corrosion by Carbon Monoxide.** M. Starczewski. (*Prace Inst. Hutn.*, (3), 133–142). Corrosion of these products by the action of CO (Bell reaction) mainly occurs in the presence of iron oxides acting as catalysts. Siliceous products, manufactured with a ferrous mineralizer (e.g. slag) show appreciably less resistance to this action, and this effect increases considerably with increased Fe content, especially over 1·25% FeO. Contents of over 1·5% cause a serious drop in resistance, and the consequent sudden fall in mechanical strength may be very dangerous in practice, where such products are used for coke oven linings. Although elevated temperatures produce a reverse reaction, C infiltration and disintegration of the structure may likewise occur, especially if the process is completed below 1000° C, and there is a drop in temperature towards the outside of the wall. This is particularly marked in a coke oven battery producing iron coke, where disintegration of the structure is highly accelerated. In the experiments, however, a mineralizer in the form of iron sulphate had no catalytic action on CO corrosion, which may be the result of ideal dissemination of the iron particles in the siliceous mixture, and at the same time by reason of greater affinity, for the formation of compounds with the silica. It is thus necessary to limit to the greatest possible extent the FeO content and to carry out the manufacturing process so that all the iron is converted into silicide.—M. T.

**Occlusion Experiments on Synthetic Mullite and Substitution Experiments with Gallium Oxide and Germanium Dioxide. Part II.** G. Gelsdorf, H. Müller-Hesse and H.-E. Schwiete. (*Arch. Eisenh.*, 1958, **29**, Aug., 513–519). Occlusion experiments are described with the oxides of Al, Fe, Cr, Ti, Si, Be, V and B, and the alkalies and alkaline earths in mullite at temperatures up to 1700° C, and the results discussed. Mullites rich in oxide, the zone of resistance of mullite in the binary system $Al_2O_3$–$SiO_2$, and the production of isotopic compounds of mullite with gallium oxide and/or germanium dioxide are also discussed.

**The Manufacture of Dolomite Bottoms at the Forges of the Providence Works of Marchienne.** M. Ledune. (*Rev. Univ. Min.*, 1958, **101**, Aug., 280–283). The author describes the new dolomite plant at Marchienne-au-Pont, and gives an account of the new principles and techniques involved.

## IRON AND STEEL, GENERAL

**The Present Position of the Swedish Iron and Steel Industry.** (*Jernkonts. Ann.*, 1958, **142**, 217–228). The production of iron and steel in Sweden over a number of years is compared with that of other countries, and an analysis of Swedish production of iron, steel and finished and semi-finished products is given.

**Expansion of Steel Production Capacity at Iscor.** (*Engineer*, 1958, **206**, Aug. 29, 347–348).

**Ferrous Metallurgy in Poland.** K. Zemaitis. (*Hutnik*, 1958, **25**, Jan-Feb., 1–5). [In Polish]. A review of the development of the iron industry after the last war is given. With the help of Russia the industry was rebuilt in 1947–1949 and a complete reorganization was carried out in 1950–1957. The production of steel increased from 43 kg per citizen in 1937 to 94 kg in 1949 and 195 kg in 1956. In 1957 the overall production of steel was 5,400,000 t.

**Progress at Durgapur Steelworks.** (*Overseas Eng.*, 1958, **32**, Sept., 48–49). An illustrated account.

**The Growing Canadian Steel Industry.** D. S. Holbrook. (*Iron Steel Eng.*, 1958, **35**, July, 133–136). An address given at the General Meeting of the AISI in 1958. The author describes the development of the Canadian steel industry, from the standpoints of production and financial policy.

**American Steelworks Invest a Milliard Dollars in New Installations and Equipment.** (*Met. Constr. Mechan.*, 1958, **90**, June, 443–445). Based on a report by the A.I.S.I., an account is given of development plans in the iron and steel industry of U.S.A.

**Republic in the South.** T. J. Ess. (*Iron Steel Eng.*, 1958, **35**, April, R.1–R.23). The article describes in some detail the Alabama plant of Republic Steel Corp., which has annual capacities of 565,000 t of coke, 525,000 t of iron, 789,000 t of O.H. steel and 408,000 t of electric steel and a wide variety of rolled products including sheet, bars, wire rod, plates, C.R. sheets, electric fusion weld pipe, nuts and bolts, etc. Data are given on the coke-ovens, blast furnaces, O.H. and electric steelmaking plants, blooming mill, bar mill, rod mill, wire mill, plate and strip mills, pipe mills, finishing departments and services.—M. D. J. B.

**Perspectives for the Brazilian Iron and Steel Industry.** J. G. Haenel. (*Eng. Min. e Met.*, 1958, **27**, 259–262). [In Spanish]. World production and economics are discussed with particular reference to their effects in Brazil. The development of the Brazilian industry is described.—P. S.

## BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

**High Temperature Technique and the Blast Furnace.** J. Wotschke. (*V.D.I.Z.*, 1958, **100**, June 11, 725–735). The desirability of carrying out reactions at the highest feasible temperature, and the need to transform the normal type of shaft furnace into a combustion chamber surrounded by thin reactive layers are used as the basis for suggesting modifications in blast furnace design.

**On the Properties at Elevated Temperatures of Blast Furnace Bricks. II. Measurements of Gas Permeability at High Temperatures.** T. Hayashi. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1032–1033). The apparatus is described, and gas-permeability figures are given between 0° and 800° C for imported and domestic bricks and sillimanite mortar.—K. E. J.

**On the Repairing of Kukioka No. 2 Blast Furnace and its Blowing-in.** T. Yamamoto. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 935–936).

**On the Filling and Blowing-in of Nakayama No. 1. Blast Furnace.** A. Ikemoto. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 933–934). [In Japanese]. The distribution of the charge throughout the furnace at filling is shown; the furnace has an auxiliary distributor and the pattern is compared with that of No. 2 furnace which has not.—K. E. J.

**Improvements in Blast Furnace Charging Gear.** L. Ya. Matusevich. (*Metallurg*, 1958, (2), 5–7). An improved top charging gear at Kuznetsk is described. Instead of six rollers, the distributor is provided with three blocks of rollers in roller bearings which ensure five years' operations without repairs. Other features of the top gear are described. A saving of about £20,000 per year was achieved due to the reduction in repairs and elimination of leakage.—B. S.

**Rational Charging Gear for Blast Furnaces.** Yu. A. Popov, K. G. Umrikhin and A. K. Shaposhnikov. (*Stal'*, 1958, (1), 7–14). The gear developed at Chelyabinsk is described and its advantages are noted. A model furnace one-tenth full size was used for investigations. Uniform distribution of materials related to lump size and composition, constant maintenance of radial distribution, absence of a rotary distributor and of distributing panels are attained and furnace control is simplified and can be made automatic. The design is for 1386 m³ furnaces and would be even better for 2000 m³ furnaces.—B. S.

**Carbon Blocks for Blast Furnace Linings.** A. N. Red'ko. (*Metallurg*, 1957, (1), 7–10). When blast furnace No. 1 at Novo Tula was driven at the rate of 1·5 t of coke per m³ per 24 h with poorly prepared burden, there was considerable wear of the refractories and scaffolding occurred. In May, 1952, 12 courses of carbon blocks 34 in. long and 15 in. thick were installed half way up the stack. Total height of the 12 courses was 11 ft 6 in. Oxygen-enriched (27–29%) blast increased the smelting rate but, with badly prepared charge, peripheral working occurred, thus increasing lining wear. The furnace produced foundry iron for 18 months and O.H. iron for 22 months from May 1952 to September 1955. Scaffolding occurred six times and was removed by explosives. It normally formed on the fireclay lining but once it extended over the carbon blocks also. Since 1955, this furnace has been producing ferromanganese. During the first seven months on FeMn, the fireclay lining above the carbon blocks was worn back 28 in. Lining wear was determined by inserting thin tubes through the stack walls. These tubes protruded 6 ft inside the furnace. As the charge descended, the tubes bent and the distance between the bend and the furnace shell indicated the degree of lining wear or the extent of scaffolding. The carbon blocks in the stack exhibited good wear resistance for five years and Russian opinion was that they should be used more extensively in the stack.—B. S.

**Hot-Cooling of Blast Furnace Tuyeres.** B. I. Ragin. (*Metallurg*, 1958, (1), 10). Because of low grade industrial water, inadequate tuyere life and loss of heat to the cooling water, the Magnitogorsk plant successfully changed over to hot cooling on all 16 tuyeres. The system is described.

**The Influence of the Mechanical Preparation of Limestone on the Production from Blast Furnaces.** A. D. Stănescu. (*Met. Constr. Masini*, 1957, **9**, (8), 9–14). [In Rumanian]. The conclusions reached are that the limestone should be crushed into lumps between 30 and 70 mm dia. and that the advantages accruing from limestone of this size are such as to outweigh the costs of crushing by manual labour where machines are not available.

**Permeability of Blast Furnace Burdens.** N. Kikuchi. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 937–938).—K. E. J.

**Measurement of Travelling Time of Blast Furnace Burden with Cobalt-60.** T. Yatsuzuka (*Tetsu-to-Hagane*, 1957, **43**, Sept., 873–874). [In Japanese]. Experimental results show a transit time of burden of 10 h, determined by $^{60}$Co tracer. The pattern of radioactivity in the iron from the start to the end of tapping is shown.—K. E. J.

**Ferro-Coke and its Use in the Blast Furnace.** D. S. Kashchenko and S. A. Sazonov. (*Koks i Khimiya*, 1956, June, 21–26). The ferro-coke produced in Russia during 1934–1937 and later and its use in blast furnaces is reviewed with comparisons with German and U.S. reports. Effects on coke-oven walls, effects of adding iron ore, especially on sulphur retention, mechanical strength of the product and its improvement, the need for dry quenching and the outlook for further advances are discussed.

**Production of Low-Manganese Pig Iron Based on Krivoi Rog Ores and Donets Coke.** P. G. Glazkov, N. E. Dunaev, A. G. Kuzub and G. A. Panev. (*Stal'*, 1958, (1), 14–20). The technical and economic effects of going over to low-Mn pig-iron production depend on the richness of the ore, volume of slag, consumption of additions, Mn content of the iron and the slag basicity. At the Stalinsk

plant, productivity increased 5–6%, coke consumption fell by 5–6% and the consumption of Mn ore by 73·5% with an increase in flux consumption of 6·72%. Efficiency of Mn utilization was doubled and cost of pig iron was reduced 5·18%. Aluminous slag is advantageous and addition of finely ground $Al_2O_3$ to the sinter may be necessary.

**Automatic Control of Heating in Hot Blast Stoves.** L. V. Brodetskii and P. G. Pikalov. (**Stal'*, 1958, (1), 21–22). A scheme using all available fan power is necessary so that gas and air consumption are related and heating can be carried out with maximum possible consumption of gas. Isodromic regulators are not essential, but control can be effected by means of electronic potentiometers with built-in contact.—R. S.

**Gas Turbines for Process Applications.** D. F. Bruce. (*Blast. Fur. Steel Plant*, 1958, **46**, Feb., 212–214, 216). Uses as gasifiers, for supply or extraction of compressed air, and as power sources are discussed. Use for blast furnace blowing is noted.

**World's Largest Steam Driven Turbo-Blower.** Appleby-Frodingham Steel Co. *Iron Coal Trades Rev.*, 1958, **176**, Feb. 21, 451–453). The installation at Scunthorpe for blowing four blast furnaces is described.

**Effect of the Closed-top Pig Ladle.** K. Morita. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 947–948). Comparisons are given between open- and closed-top practice for pig cooling rates, the inside wall temperature of the empty ladle, and the refractory repairs needed.—K. E. J.

**Stepping Up Blast Furnace Production Still Further. Conference of Ukrainian Blast Furnacemen.** (*Metallurg*, 1958, (1), 1–2). At the Ukrainian Blast Furnace Conference in November, 1957, it was emphasized that, in spite of high top pressure operations, blast furnaces were not being driven sufficiently hard. The blast volume should be increased when high top pressure was used without necessarily increasing coke consumption. Other factors to improve blast furnace productivity were: improvement in top charging gear, more precise slag control, an increase in the ore charges, desulphurization outside the furnace, the use of natural gas for blowing the furnace. The upthrust of the gas streams in the stack must be increased by increasing pressure in the furnace chamber. Sinter must be block charged. A leading Russian blast furnaceman, N. I. Krasavtsev, stated that American and some Russian blast furnacemen who believe that higher rates of smelting must be accompanied by a rise in coke consumption are wrong.—R. S.

**On the Reduction of Hematite Ore by Solid Carbon.** K. Shimanaka. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1017). In air, the curve of reduction begins to rise steeply at 900° C, flattening out, at approx. 95% reduction at 1100° C. *In vacuo*, reduction commences at 1000° C, and at 1200° C has only reached 10%. The percentage of Si in the reduced hematite is shown for various temp.: it is far higher for vacuum- than air-reduction.

**The Reduction of Lump Ores by Bosh Gas.** Y. Takahashi. (*Tetsu-to-Hagane*, 1957, **43**, 1015–1017). Reduction-time curves are given for 13 samples (including hematite ores, magnetite ores and sinters) in a synthetic "bosh gas," and comparative reducibilities are shown graphically.—K. E. J.

**Studies on the Effect of Blast-Furnace Operating Rate and Other Factors on Coke Rate.** T. Kitagawa. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 945–947). Statistical data are given for the relationship between coke rate and output under conditions of (*a*) ash content in the coke between 8 and 12%, and (*b*) average iron content in the ore between 56 and 60%.—K. E. J.

**Study on the Distribution of Gas in a Blast Furnace. I. A Radioactive Technique for Determining Gas Transit Time in a Blast Furnace.** Y. Takahashi. **II. Determination of Transit Times Using Radon Tracer in a Driving 1000-t Blast Furnace.** K. Miyagawa. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 938–940, 941–943). [In Japanese]. I. An apparatus is described for repeating the work in which the gas transit time from tuyere to stockline is found by detecting a radioactive pulse (radon); using a pulsed ionisation chamber, a smaller pulse can be used than with a Geiger counter. II. Results are given for times at wall and centre for normal operation, slagging and tapping.—K. E. J.

## PROPERTIES, TREATMENT, AND USE OF SLAGS

**Micrographic Study of the Solidification of Blast-Furnace Slags.** L. Pardon. (*Silikáty*, 1958, **2**, (2), 176–177). [In Czech]. A series of micrographs taken on small samples of slag which had cooled at various rates, and were thus preserved at various stages of devitrification, are discussed.—P. F.

**Studies on Ferro-Manganese Slags. II. On the Apparent Equilibrium of the Si-Reducing Reaction. I.** I. Tanabe, H. Saito, T. Nishida and K. Kosaka. (*Nippon Kinzoku*, 1956, **20**, Apr., 188–191). An equation is derived for the apparent equilibrium constant $K^a$ Mn–Si at 1510° C from measurements on synthetic slags; MgO was regarded as basic and $Al_2O_3$ as neutral in calculating basicities.

**Some Problems Connected with the Utilisation of Blast Furnace Slag in the Production of Cement.** T. Rogozinski. (*Hutnik*, 1958, **25**, Jan.-Feb., 5–9). [In Polish]. Following the practice in various countries, Poland has decided to increase the utilization of slag and to extend her cement industry. A review of research carried out abroad and in Poland on the hydraulic properties of slags is given.

**Slag–Iron Blast-Furnace.** W. W. Key. (*U.S. Bur. Mines Min. Yearbook*, 1955, **1**, 1021–1029). Statistics of the output of processed slag and its uses, price, and technology are given.

## PRODUCTION OF WROUGHT IRON

**New Wrought Iron Developed by A. M. Byers Co.** (*Iron Steel Eng.*, 1958, **35**, March, 157). Information is given on a new type of wrought iron which it is claimed has better properties than standard wrought iron.

## DIRECT PROCESSES

**New Process Achieves Direct Reduction of Iron Ore.** (*Can. Min. J.*, 1958, **79**, Aug., 78–80). The Strategic-Udy process uses a rotary kiln of the type used for low-grade Mn ores. Fluxes and a low-cost fuel are fed and discharged as a free-flowing system of particles at 1100–1300° C. Electric furnace gas can be fed when little additional solid fuel is necessary. The hot charge is fed directly to an arc furnace.

**New Progress Achieved in the Direct Reduction of Iron Ores.** J. Astier. (*Min. y Met.*, 1958, **18**, July, 25–27.). [In Spanish]. Modern techniques and developments are described. The author puts the important new achievements as, (i) the first laboratory studies of very high temperature reduction by gases, (ii) the start of the hydrogen reduction process for rich and pure ores on a semi-industrial scale, (iii) the improvements in the Krupp-Renn process, and (iv) the improvements in the low shaft electric furnaces.—P. S.

**Considerations on the Direct Extraction of Iron.** E. Houdremont. (*Techn. Mitt. Krupp*, 1958, **16**, June, 24–26). Various aspects of the "direct" reduction of iron by the Krupp–Renn and other processes are considered.

**The Reduction of Iron by the Renn Process.** E. Schürmann. (*Stahl Eisen*, 1958, **78**, Sept. 18, 1297–1308). This a comprehensive review on the Renn process and its theory, dealing in particular with the physico-chemical principles involved in the reduction of iron and of the various accompanying elements from their oxide mixtures by $CO + CO_2$. The reduction of P and its effect on the carburization of the reduced iron and on the hardenability of the Renn iron are explained by means of the Fe–P–C equilibrium diagram.

**The Krupp–Renn Process in the Techniques of Beneficiation and Metallurgical Procedure.** R. Gerlach. (*Techn. Mitt. Krupp*, 1958, **16**, June, 27–33). The development of the Krupp–Renn process is outlined, and its advantages and range of application are discussed.—L. D. H.

**Reconstruction and First Operational Results of the Krupp–Renn Plant in Salzgitter-Watenstedt.** D. Fastje. (*Techn. Mitt. Krupp*, 1958, **16**, June, 34–38). The Watenstedt Krupp–Renn plant was recently reconstructed, three larger furnaces being installed, two with a daily capacity of 610 metric t, the third of 780-t capacity. The preparation and supply of the burden were also modified. Early operational results are described.

**Experimental Results of the Production of Spongy Iron in the Krupp-Renn Drum.** S. Henkel and H. Serbent. (*Techn. Mitt. Krupp*, 1958, **16**, June, 41–46). Trials carried out in a pilot Krupp–Renn plant are described, in which spongy iron was produced in such a way that re-oxidation of the iron was practically eliminated. Results are reported on a number of ores and reducing agents.—L. D. H.

**The Behaviour of Various Slag Compositions in Practice and Their Influence on Production.** W. Heerwagen. (*Neue Hütte*, 1958, **3**, June, 351–357). Slag practice in the Krupp–Renn process at VEB Maxhütte, changes in composition from 1954–1957, and the influence of various constituents are discussed.

## PRODUCTION OF STEEL

**The Dunes Works of the Société des Forges et Aciéries de Firminy.** (*Aciers Fins Spéc.*, 1958, July, 96–97). A short description of the development and production of this works.

**Activity and Preoccupations of the French High-Grade Steelmaking Industry.** (*J. Four Elect.*, 1958, **63**, (2), 65–66). Economic and other aspects of recent French production of high-grade and special steels are discussed.

**Power Layouts in Steel Mills.** A. J. Mosso. (*Iron Steel Eng.*, 1958, **35**, Feb., 111–119). All aspects of power layout systems are considered. A balance must be struck between cost and reliability of equipment. The layout should be planned to reduce maintenance problems, provide access to components, and ensure safety of personnel.—K. E. J.

**The Hydrocyclone as Part of the Water Economy of an Integrated Iron and Steel Works.** A. Köhler. (**Stahl u. Eisen*, 1958, **78**, Feb. 20, 235–239). Water economy of iron and steel works is discussed and various designs of hydrocyclones and their efficiency are described.—T. G.

**Steelmaking German–American Style.** (*Steel*, 1958, **142**, Mar. 10, 130, 133). A brief account of the rebuilt August Thyssen-Hutte works of Duisburg.

**Tula, Sverdlovsk, Chelyabinsk, Magnitogorsk: Key Russian Steel Plants.** N. J. Grant and J. Chipman. (*J. Met.*, 1958, **10**, Mar., 182–185). A general account of a visit.

**The Imphy Metallurgical Company Pamiers Works.** (*Aciers Fins Spec.*, 1957, Dec., 127–131). This paper summarizes the history and development of these works. Carbon and alloy steels are supplied as forgings, pressings, castings and rolled products. The lay-out includes a foundry, rolling mill, machine shop, heat treatment shop, centrifugal casting and a shop for light alloys.—R. P.

**Republic Steel Corporation Expands Cleveland Plant to 3,360,000 tons Capacity.** (*Blast Furn. Steel Plant*, 1958, **46**, Jan., 60–62). A short description of new and modernized plant recently installed at this works is given. This includes: two new 375-t O.H. furnaces and expansion of four furnaces from 275 to 375 t, a new 45 in. universal slabbing mill, 16 new soaking pits and a new coke oven battery.—B. G. B.

**Exchange Current Between Liquid Metal and Slag.** Yu. P. Nikitin and O. A. Esin. (*Doklady A.N.*, 1957, **116**, (1), 63–65). The experiments were conducted in a specially constructed MgO crucible, at temperatures 1480–1500° C in a carbon resistance furnace. The electrodes carrying a.c. were of liquid metal (Fe–C and Mn–Si alloys). It is shown that the potential at the slag/metal interface is created by the passage of Fe ions, from the metal into the slag. It is also shown that

there is a linear relation between the surface and the interphase tensions and the composition of the metal.—S. I. T.

**Efficiency and Outlook for the Development of Basic Bessemer Steel Production in the U.S.S.R.** V. M. Tsitver. (*Stal'*, 1958, (2), 164–168). A general review of the process compared with the O.H. process concluding that the Kerchensk and Lissakovskoye deposits are suitable for the process and would support a production of 10–12 × $10^6$ t/year. Gas producers and liquid fuel plants would not be needed and labour productivity is on the average 25% higher than that of tilting O.H. furnaces. Costs are 4·5–5% lower and capital costs 40–50% lower than for O.H. shops.

**Some Technological Comments on the Construction of Acid Converters in Steelworks.** Z. Zdenek and F. Prochaska. (*Hutnik*, 1958, **8**, March, 85–87). The problem of the installation of such converters or the conversion of existing plant, and refractory problems are discussed. A comparison is made between the simple production of conventional types of crude O.H. steel of average quality (0·25% C max. rimming and killed steels), and the duplex process with acid converters and electric arc furnaces. The latter alternative has many advantages, notably greater output and the production of qualities of steel required for forging.—M. T.

**The Decarburization Reactions in the Basic Bessemer Converter During the Dephosphorization Period.** H. Pottgiesser jun. (**Stahl u. Eisen*, 1958, **78**, Mar. 6, 291–298). The equilibrium of carbon and oxygen in blowing with air, oxygen-enriched air, a mixture of oxygen and carbon dioxide, and nitrogen-enriched blast were studied. The effect of nitrogen on deoxidation is discussed. Using nitrogen-enriched blast, steels of very low carbon and oxygen contents can be produced.

**Hydrodynamic and Practical Examination of the Bath Movement in the Bottom-Blown Converter.** P. Leroy and G. Cohen de Lara. (**Rev. Mét.*, 1958, **55**, Jan., 75–97; Feb., 186–200). Hydrodynamic study of the movement of the bath has enabled a systematic examination of the geometrical and pneumatic factors affecting the 'blowability' to be made. A theory to explain the oscillations occurring in the bath has been developed. From the results obtained it is shown that it is necessary to provide for as wide a blast zone as possible and hence the peripheral tuyeres should be as close as possible to the inner boundary of the lining.—B. G. B.

**Working Results in the Blowing of Basic Bessemer Steel Using the IRSID-Opacimeter.** H. vom Ende and G. Mahn. (**Stahl u. Eisen*, 1958, **78**, Apr. 3, 412–418). The authors describe the opacimeter and the modifications necessary for practical use. The use of the instrument becomes more and more difficult when the temperature of the bath increases. The authors doubt whether the instrument will ever be a reliable tool for the production of steel. In the following discussion, very interesting points are raised particularly from IRSID members.—T. G.

**Refining of Phosphoric Iron and in Particular, Thomas Iron with Pure Oxygen by the I.R.S.I.D. Process.** B. Trentini and M. Allard. (*Centre Doc. Sidér. Circ.*, 1957, **4**, (11), 2259–2266). This process consists of directing a jet of oxygen on to the surface of the liquid iron and suspending lime and fine ore in the gas stream. The heat evolution raises the temperature to 1600° C and the slag remains liquid and after 12 to 15 mins contains 20% $P_2O_5$. After deslagging, further addition of oxygen and lime is made. Full details of results obtained on a 3-t scale are given. One of the major advantages of the technique is the facility of removing P without the need to reduce the metal-carbon. Very low $N_2$ contents are found. 70% desulphurization also takes place during refining. Tests carried out on steel made by this new method show excellent deep-drawing properties.—B. G. B.

**Burner for Firing the Bottoms of Converters.** G. de Rousiers. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **4**, (12), 2425–2429). A burner is described for use with blast furnace gas and oxygen. The burner is designed for placing inside converters and is cheaper in operation than the conventional practice using coke.—B. G. B.

**Desulphurization in the Basic Bessemer Converter.** K. G. Speith, H. vom Ende, and H. Vosskötter. (**Stahl u. Eisen*, 1958, **78**, Feb. 6, 152–156). Analysis of production heats showed that desulphurization depends more on the composition of the slag than on the temperature of the heat. The better desulphurization of oxygen-blown heats is attributed to an earlier liquefaction of the slag than in conventional blowing. The higher the iron content of the slag the lower is the desulphurization rate. Desulphurization via the gas phase was not found.—T. G.

**The Use of Magnesium in the Production of Converter Steel.** K. S. Prosvirin, V. I. Baptizmanskii, M. P. Kuznetsov and V. D. Umnov. (*Metallurg*, 1957, (1), 16–17). The use of silico–magnesium (64%Si, 11·8%Mg) instead of 45% ferrosilicon for the deoxidation of Bessemer rail steel was tested in 10 heats. Mechanical properties of the experimental metal were found to be satisfactory and the impact strength of the metal was similar to that produced in O.H. furnaces. The use of Mg considerably decreased the proportion of non-metallic inclusions (from 0·0427% to 0·026%) and decreased the grain size and sulphur content of the metal.—V. G.

**Results of Tests on Thomas B.T.M. Steel Made with Oxygen-Enriched Air.** M. Thiault and J. Stremsdoerfer. (*Pub. Inst. Rech. Sidér.*, 1957, series A, No. 170, Sept., 1–107). The results of a large number of tests carried out on steel made by the oxygen enriched Bessemer converter are compared with steel made with the normal air blast, oxy-steam blast and in an O.H. furnace. The steel was rolled into sheet form and a wide range of tests made including deep drawing and enamelling. Wire drawing tests were also carried out. The results show that the oxy-air converter can make steel having deep drawing properties far superior to steel from the normal air converter and equivalent to O.H. steel.—B. G. B.

**Improvement of the Metal Quality as an Important Possibility of Increasing the Output of Open Hearth Furnaces.** N. I. Lopukhov. (*Metallurg*, 1956, (5), 9–11). Operating practice of the O.H. melting shop of the Magnitogorsk Metallurgical Combine is outlined. It is pointed out that by improvements in the operating practice the proportion of metal passed into scrap can be decreased thus increasing the overall output of furnaces.—V. G.

**Intensication of Steel Making by the Use of Oxygen.** G. A. Podol'skaya and V. A. Makovskii. (*Metallurg*, 1956, (6), 17–19). The use of oxygen for the intensification of the steel-making process in tilting 350 t. O.H. furnaces operating with 75% pig iron (phosphorus contents up to 1·7%) at the "Azovstal'" Works is outlined. Oxygen is introduced into the flame through a specially made water cooled tuyere and into the bath through a 1¼ in. tube passed through specially made holes in the back wall (diagrams are given). Comparison of the results obtained with and without oxygen is made. It is pointed out that at the present low availability of oxygen, the overall output increased by 4 to 5%. The quality of rimming and rail steel is not affected by the use of oxygen.—V. G.

**Four-Period Scheme of Combined Automatic Control of Thermal Conditions in an Oil-Fired Open Hearth Furnace.** V. P. Borodin, P. E. Darmanyan, A. A. Yudson and A. V. Vasil'ev. (**Stal'*, 1958, (2), 114–120). Various conditions to be observed for satisfactory operation are given, and fuel consumptions for the production of carbon and Cr steels are shown.—R. S.

**The Influence of Alumina and Magnesium Oxide on the Desulphurization of Pig Iron Melts with Lime–Silica Slags under Reducing Conditions.** W. Oelsen, E. Schürmann and S. Osman. (*Arch. Eisenhüt.*, 1958, **29**, Apr., 205–218). After referring to the results of previous work, the authors describe similar trials, carried out to determine the influence of increasing $Al_2O_3$ and/or MgO additions to lime–silica slags with given $CaO/SiO_2$ ratios. Although $Al_2O_3$ and MgO, separately or together, improve desulphurization, silicon reduction is also influenced.

**The Control of Open-Hearth Furnaces and their Peculiarities when Unpurified Producer Gas is Used as Fuel.** G. Oppelt. (*Neue Hütte*, 1958, **3**, Jan., 26–31). The metallurgical process is not yet susceptible to automation; the thermal process is automatically controllable. The use of unpurified producer gas as fuel necessitated the development of a special gas–air ratio control.—L. J. L.

**Contribution to Deoxidation Processes.** E. Plöckinger. (*Berg-Hütten. Monatsh.*, 1957, **102**, Nov., 277–283). The author discusses deoxidation processes in steelmaking, including the changes in oxygen concentration in liquid steel during refining, carburizing and deoxidation with Si, Al etc. He considers that oxide precipitation is influenced by surface tension and wetting properties. The uses of steels depend partly on the type, shape and plastic behaviour of the oxides formed. By controlling deoxidation, it should be possible to obtain oxide inclusions with pre-determined properties. (12 references).—R. P.

**Industrial Testing of Open-Hearth Ports with Ejection of Hot Air.** N. I. Kokarev, V. G. Lisienko, Ya. A. Goncharevskii and V. G. Beloshapkin. (*Metallurg*, 1957, (12, 28–29). Model and full-scale experiments on a system of hot-air ejection into the gas port of a mixed (CO and B.F.) gas fired O.H. furnace are described. The end used was designed by the Ural Polytechnic Institute and incorporates a by-pass for distributing the combustion products more effectively between the gas and air checkers. In spite of some design and operating defects the new end led to the consumption of standard fuel falling to 110–115 kg/tonne although a higher than normal firing rate (up to 33–34 × $10^6$ Cal/h) was adopted during charging.—S. K.

**Some Observations on the Applicability and the Economics of Automatic Operation of Open-Hearth Furnaces in Industrial Practice.** H. Klepsch. (*Neue Hütte*, 1958, **3**, Jan., 20–26). Automatic control of O.H. furnaces has come to stay. Some general observations are therefore presented to enable founders to apply the method correctly, with reference particularly to furnace pressure regulation and gas–air ratio control.—L. J. L.

**The Basic Roof—Extended Hearth—Use of Oxygen—Firing Rates at Granite City Steel Company.** A. L. Howard. (*Blast Furn. Steel Plant*, 1958, **46**, Jan., 49–51). A brief description of the change-over from all silica to basic roofs at this works is given. Fuel costs have been decreased and furnace outputs increased.

**Possibilities of Increased Steel Production in Open Hearth Furnaces Through a Suitable Analysis of Technical and Economic Indicators.** P. Kielski. (*Hutnik*, 1957, **24**, April, 153–158). [In Polish]. Methods of evaluation of different technical and economic indicators per t of steel produced, are applied to the analysis of the results obtained in the O.H. furnace. This was done in order to increase the impartiality and accuracy of the evaluations.

**Studies on the Decarburisation Reaction of Molten Fe–C Alloys. IV. On the Decarburisation Reaction with Water Vapour.** Y. Katsufuji. (*Tetsu-to-Hagane*, 1957, **43**, Nov., 1205–1210). [In Japanese]. Carbon removal was studied in molten Fe–C alloys containing 0·5–4·0% C in contact with $H_2$–$H_2O$ and $O_2$–$H_2O$ gas mixtures. In the former case, the rates of carbon removal were proportional to the partial pressure of water vapour in the range 7–32 mm Hg. Increased amounts of carbon removal were found in the latter case, depending on the oxygen gas pressure.

**The Problem of Steel Desulphurisation in the Basic Open Hearth Furnace.** T. Mazanek. (*Hutnik*, 1957, **24**, Feb., 55–58). [In Polish]. The author discusses the up-to-date views on desulphurization of steel in the basic O.H. furnace. In order to increase the rate of desulphurization he recommends: (*a*) composition of the charge after melting should yield slag of 2·0% basicity and contain 20% of FeO; (*b*) Removal of the initial slag and

accelerated formation of the slag of higher basicity and higher FeO content.

**Studies on the Desulphurization Reaction of Molten Iron with Gas. III. Interaction Coefficients of Sulphur in Molten Iron with Several Components and Specific Rate Constants of Desulphurization.** Y. Katsufuji. (*Tetsu-to-Hagane*, 1957, **43**, Dec., 1281–1285). [In Japanese]. Reaction rates were measured for desulphurization by free $H_2$ and $O_2$ of molten Fe–S–X and Fe–C–S–X alloys (X = C, Si, Al, Ni, Cr or Mn). The reaction rate was increased by addition of C, Si, or Al; decreased by Mn; and unaffected by Ni or Cr. The effects of the elements added corresponded with the activity of sulphur in the melt. A relationship is given between the specific rate constant, the activity coeff. of sulphur and the interaction coeff. of sulphur.—K. E. J.

**The Reactions of Carbon and Sulphur and Metal/Slag Equilibrium in the Refining of Steel with Oxidizing Slag.** G. Violi. (*Met. Ital.*, 1957, **40**, Dec., 805–816). [In Italian]. The contribution which physical chemistry and thermodynamics have made to the understanding of the equilibrium of the reactions between metal and slag in steelmaking is briefly discussed. Oxygen content in steel and the carbon/oxygen equilibrium are studied. The distribution of $O_2$ in the slag and in the pure iron is examined as well as the kinetics of decarburization and the metal/slag equilibrium. The means of controlling the decarburizing reaction is described. The desulphurization of steel with oxidizing slag and the distribution of S in steel are discussed. Technical and statistical data are given as well as tables and groups. (23 references).

**Reduction of Lost Time of Open Hearth Furnaces for Cold and Hot Repairs.** T. G. Shunin and A. I. Korolev. (*Metallurg*, 1956, (5), 14–17). Organization of cold and hot repairs to O.H. furnaces of the Magnitogorsk Metallurgical Combine is described.—V. G.

**An Improvement in the Durability of Open Hearth Furnaces.** Ya. G. Privezentsev and V. Ya. Ignatenko. (*Metallurg*, 1956, (1), 18). Measures taken in order to improve the service life of basic O.H. furnaces are outlined. The main points: top four courses of regenerators made from chrome–magnesite bricks, the rest from chamotte bricks; the back wall (above the slag zone) was lined with chrome–magnesite bricks with metallic interlocks.

**Fettling Machine for Open Hearth Furnaces.** F. F. Svirdenko and D. K. Kharlamov. (*Metallurg*, 1957, (1), 35). A description of a fettling machine operating by spraying fettling materials with compressed air is given.—V. G.

**Mixture for Fettling Open-Hearth Furnace Bottoms.** I. V. Panyushkin and M. K. Musikhina. (*Metallurg*, 1957, (12), 20). A brief account is given of conditions for the successful use of a three component (80% magnesite powder, 13% 0–4 mm raw dolomite and 7% scale) for fettling O.H. furnace bottoms in a duplex shop.—S. K.

**Bulldozer for Clearing Slag Pockets.** I. I. Lesunov and V. E. Pis'mennikov. (*Metallurg*, 1957, (12), 21–22). A brief account is given of the use of a modified tractor for the rapid (3–12 hr) clearing of O.H. furnace slag pockets. Before the adoption of this device the time required was 1·5–2 days.—S. K.

**Clearing Slag Pockets with the Aid of Removable Plates.** G. M. Kroitman. (*Metallurg*, 1957, (12), 22–23). A very brief account is given of a practice in which the O.H. slag pockets have old chrome–magnesite brick false walls and bottom supported on metal plates resting on cast-iron balls. To clear the pocket the whole slag-filled structure is drawn out with the aid of a pouring crane.—S. K.

**Use of Removable Boxes for Removing Slag.** V. B. Kaplun and E. D. Akol'tsev. (*Metallurg*, 1957, (12), 23–24). A very brief account is given of O.H. slag-pocket practice in which a wheeled and refractory-lined box collects the slag and is removed periodically. The slag pockets are provided with cooled reinforcements.—S. K.

**Metallic Recuperators.** A. Stojek. (*Hutnik*, 1957, **24**, Feb., 58–61). [In Polish]. Progress in metallic recuperators, especially methods of lowering the wall temperature, the construction and the mounting of heating elements, allowing for their movement with the change of temperature, are discussed.

**The Checkers in Open Hearth Furnaces.** L. Halm. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **4**, (12), 2465–2479). Factors influencing heat transfer between gas and brickwork is first considered. The influence of dust on the checker brickwork is discussed together with the types of brick available for checker construction.—B. G. B.

**The Choice of Refractories for Checkers for Open Hearth Regenerators.** S. L. Gribov and A. M Banshchikov. (*Metallurg*, 1957, (1), 19–23). The service life of basic O.H. furnaces with chrome–magnesite roofs is often limited not by the durability of the main roof, but by that of the regenerators. An investigation of the durability of a regenerator made from different refractory materials is described. The following refractories were tested: Dinas bricks, ordinary chamotte bricks, high quality chamotte bricks (40–42% $Al_2O_3$) and chrome–magnesite bricks for the top courses. The best results were obtained with regenerators made from high alumina chamotte bricks with 6 top courses of chrome–magnesite bricks.—V. G.

**The Lining of Ceramic Recuperators.** M. S. Krustal' and V. A. Pluzhnik. (*Metallurg*, 1958, (3), 28–30). A new technique used at Krivoi Rog for bricking recuperators is described. The type of shaped brick and the recuperator are illustrated.—R. S.

**Study of Special Refractories for Open Hearth Furnace Checkers.** M. Faure. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **4**, (12), 2481–2483). A brief review is given of recent experiments carried out at the Hayange plant in France. Shape of bricks and their composition have been studied.—B. G. B.

**The Temperature of the Products of Combustion on Exit from the Furnace Chamber of Open Hearth Tilting Furnaces.** G. M. Glinkov, E. A. Kapustin and V. A. Makovskii. (*Stal'*, 1958, (3), 223–224). Waste gas temperature is higher than soft temperature and is ~ 1620–1650° C. In the uptakes it is 60–80° lower than in a fixed furnace which reduces preheat of the air and furnace efficiency. Filling up of cracks to reduce entrainment is important. Measurements of temperature and rates of change were made continuously with a radiation pyrometer. Measurements were also made between reversal valves. Enrichment with 24–25% $O_2$ was also studied.—R. S.

**An Intersection Point Method as an Aid to the Calculation of Heat Transfer.** K. Winkler. (*Radex Rund.*, 1958, (2), 63–71). The application of the graphical method of solving the standard equation to recuperators is included.

**A Statistical Analysis of the Technology of Tube Steel Production.** M. Knotek. (*Hutn. Listy*, 1958, **13**, (3), 287–298). [In Czech]. Factors influencing reject rates in tube steel production, such as: (*a*) personnel, e.g. the open-hearth team-leader, (*b*) initial and final sulphur contents of the melt, (*c*) Mn : S ratio, (*d*) period of pure boil of the melt, (*e*) amount of Al additions, and (*f*) pouring rate were used in a statistical analysis to determine the main causes of the occurrence of rejects. Mathematical methods used, the calculating machines employed and the significance of the correlation analyses are discussed. The results, based on about 170 melts, clearly showed the importance of the effective control of the length of the period of dephosphorization of the melt. The carbon content in the final product was shown to be a significant factor in relation to the mechanical properties of the tubes, but no relation was found between the mechanical properties of the as-cast metal and the metal as embodied in the finished tubes. Obviously, the process of manufacture of the tubes was here decisive. Further statistical work of a similar nature is being planned.—P. F.

**Comparison of Production Costs Using the Open Hearth and Electric Arc Furnaces.** J.-C. Robin. (*Mét. Constr. Mécan.*, 1958, **90**, Apr., 291–293). The advantages in costs of the electric arc furnace compared with the O.H. furnace when using a cold charge are emphasized.

**The Basic Hot-Blast Cupola, the Open-Hearth Steel Plant and the L-D Converter in Combined Operation.** A. Richter, G. Cohnen and P. Jacobi. (**Stahl u. Eisen*, 1958, **78**, Mar. 6, 273–284). To increase the steel production of the Gussstahlwerke Witten, an O.-H. and arc-furnace steel works, a basic hot-blast cupola of 15 t/h output was installed. The design of the cupola is described in detail. For duplexing with the arc furnace an L-D converter was added together with an oxygen plant of 1500 m³/h.—T. G.

**Production of Electric Steel by the Oxygen Converter-Electric Furnace Duplex Process.** S. D. Gagin. (*Kislorod*, 1957, (6), 1–11). The oxygen (top-blown) converter-electric furnace duplex process reduces capital costs by 12 to 16% and steel costs by 3 to 7%. It eases the scrap supply position for arc furnaces operating on soild charge and doubles furnace productivity. Detailed cost and material consumption data are given.—R. S.

**The Role of the Basic Hot-Blast Cupola in an Integrated Iron and Steel Works.** H. Voigt. (**Stahl u. Eisen*, 1958, **78**, Mar. 6, 284–291). The efficiency of the O.H. steel plant at Salzgitter was markedly increased by the use of three basic hot-blast cupolas. Using the cupolas, low-grade scrap and high-acid basic-converter iron can be prepared for use in the O.H. furnaces. The plant and its mode of operation are discussed.—T. G.

**Appraisal of European Electric Furnace Melting.** R. Boutigny. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 6–20). A review of furnace type and refractories and output with brief accounts of the industry in Italy, France and Belgium and Great Britain.

**Arc Furnaces. [Furnaces for Steel-Making and Chemical Industry].** F. Beck. (*Berg-Hütten. Monatsh.*, 1957, **102**, Nov., 283–291). The temperatures obtainable with various electrodes are discussed and the conditions under which arc furnaces can be used with advantage are outlined. The construction, operation and performance of various furnaces are compared.—R. P.

**Fifty Years of Progress in Electric-Arc Furnaces.** S. Arnold. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 20–30). An illustrated review.

**Evolution of the Electric Furnace in the Manufacture of High Quality Steel.** (*J. Four Elect.*, 1957, **62**, Nov.-Dec., 205–206). An outline is given of the development of the electric furnace for steelmaking.—L. D. H.

**New Furnace for Alloy Steel Productions.** (*Brit. Steel.*, 1958, **24**, Mar., 76, 79). The introduction of a mains-frequency coreless induction furnace for the production of high-speed steel is described. Previous fears of non-metallic inclusions in the steel have proved groundless, and capital and running costs are favourable The 'Birlec' furnace was developed specifically for melting high-grade steels. It has a capacity of 1½ t at ½ t/h, with a power input of 450 kw.—D. L. C. P.

**Trends of Electric Furnace Growth as They Pertain to Arc-Furnace Transformers.** R. Bry. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 31–36). Output is first discussed statistically. Relatively little is said about transformers.

**Rapid Capital Repairs of Large-Capacity Electric Furnaces.** M. G. Dmitrienko. (*Metallurg*, 1957, (12), 26–27). For the reduction of idle time during major repairs of electric furnaces, the removal of the old furnace and its replacement with a previously lined unit is adopted at the "Dneprospetsstal'" works. A special trestle into the casting bay is constructed to enable the relatively large pouring cranes there to be used for handling the old and new furnaces. The handling procedures are outlined. The new method has enabled idle time to be reduced from 8 to 2½ days.—S. K.

**Quality Control in the Electric Furnace Shops.** J. J. Oravec. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 212–218).

**An Efficient Preheater for Electric Furnace Charges.** R. J. Leary and W. O. Philbrook. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 300–313). A top-fired preheating furnace is described. Efficiencies and costs are reviewed.

**Why We Use Oxygen in Our Acid Electric Furnace.** E. J. Etowski and R. Perry. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 314–327). A brief note on the advantages found by the Farrell-Cheek Steel Co. and extensive discussion and accounts of experience at other plants by other contributors are given.

**Virgin Melting Stocks for Steelmaking.** P. E. Cavanagh. (*A.I.M.M.E. Elec. Furn. Steel Conf. Proc.*, 1956, **14**, 281–286). Ore, liquid pig iron or blown metal, solid pig or sponge iron and H-iron are reviewed. **Practical Aspects of Using Sponge Iron in Electric-Arc Furnaces.** J. L. Stalhed. (286–293). Swedish practice using W-S sponge iron is described. **Use of Hot Metal in Electric Furnaces.** W. A. Hoff. (293–295). **Future of Desiliconized Pig Iron for the Electric Furnace.** S. L. Case. (295–299).

**New Results Obtained in the Study of Temperature Variations in Electric Furnaces as a Function of Time and Power Output.** J. Bréfort. (*Chal. et Ind.*, 1957, **38**, Dec., 325–328). A simple relation between the power, (W), and the temperature differences $(\theta-\theta_0)$ in different parts of the furnace is given by the equation $W = \alpha(\theta-\theta_0)^{\beta}$. $(\theta-\theta_0)$ is proportional to a given time. Industrial applications are discussed.—R. P.

**Methods of Obtaining Some Economy in the Consumption of Ferro-alloys During the Production of High Speed Cutting Steels.** E. I. Mashkevich. (*Metallurg*, 1956, (4), 29–31). By suitable timing of ferroalloy additions and their correct placing in an electric furnace, some economy in the consumption of ferro-alloys can be achieved. The practice of producing high speed cutting steels used in the Dnepropetrovsk Works is outlined.—V. G.

**Economics of the Energy Used in a Steel Works with Special Reference to the Use of Electric Heating.** K. Sauer. (*Electrowärme*, 1958, **16**, Mar., 71–75). The Rheinhauser Steelworks is surveyed. Blast furnace gas is used as far as possible, and coke, coal dust, fuel oil and electricity are also used. Two 120-t arc furnaces are described competing successfully with O.H. furnaces.

**Uses of Electric Heating in a Works Producing Refined Steel.** F. Hahne and H. Schmitz. (*Elektrowärme*, 1958, **16**, Mar., 62–67). Uses in the Edelstalwerke A.G. Krefeld are described. There are 13 arc furnaces with capacities from 1·5–70 t with a total consumption of 1000–1500 kVA. There are also 2 induction furnaces and a vacuum furnace for special purposes. Some resistance furnaces are used for annealing and tempering sheets and bars and for hot-working and salt-bath treatment. Of the annual consumption of $264 \times 10^6$ kWh, $191 \times 10^6$ are used for heating. The furnaces are illustrated.

**The Effect of Inductive Bath Agitation on the Metallurgical Reactions Taking Place in the Arc Furnace.** K. G. Speith, H. vom Ende and W. Schneider-Milo. (**Stahl u. Eisen*, 1958, **78**, Feb. 20, 215–220). Results of inductive agitation on the metallurgical reactions during production runs were studied in a 30-t arc furnace. During the oxidation period no effect of inductive agitation was observed as the agitation of the bath by the decarburization reaction is more vigorous than the additional agitation from the induction current. In the fining period, inductive agitation accelerates the reactions between bath and slag resulting in lower sulphur and oxygen contents at shorter fining times.—T. G.

**Balance Account for Materials and Power of a Steel Plant Comprising Electric Low-Shaft Furnaces and Arc Furnaces.** G. Schweisgut. (*Stahl u. Eisen*, 1958, **78**, Apr. 3, 407–412). A detailed account is given for the planning of a steel plant based on electric low-shaft furnaces and arc furnaces. The energy for the production of 1 t of steel is calculated as 680,329 k.cal. The various raw materials needed are considered.—T. G.

**Electric Furnace Linings.** Lefebvre. (*Centre Doc. Sidér. Circ.*, 1957, **4**, (9), 1857–1862). The properties of materials used in the construction of electric furnace linings are considered. The cost of different linings over a furnace campaign are compared. Increased lining lives and a decreased cost of lining/ton steel is obtained with magnesia and silica bricks in place of stabilized dolomite construction.

**Influence of the Number of Charges on the Melting Time.** Delbey. (*Centre Doc. Sidér. Circ.*, 1957, **4**, (9), 1853–1855). The influence of the type of scrap and amount added on the time of melting in a 12-t electric furnace is presented in tabular form.—B. G. B.

**Removal of Sulphur from Steel in the Ladle.** A. S. Tochinskii and N. V. Popova. (*Stal'*, 1958, (3), 214–218). Treating acid electric steel with liquid desulphurizing slags prepared on the metal in the furnace reduces sulphur by about 40%. When treating basic O.H. steel in the ladle the S is reduced by 35–40%; the desulphurization process is therefore independent of the type of plant. Useful tabulated data are given.—R. S.

**Deoxidation of Killed Steel in the Ladle with Ferroaluminium.** I. N. Lad'yanov. (*Stal'*, 1958, (3), 218–223). A special Fe–Al alloy is used and is superior to Al as it reduces the quantity used by half, reduces inclusions and improves ductility, gives finer grain for the same Al consumption, and the quantity used is more easily controlled.—R. S.

**Production of Steel from Low-Manganese Pig Iron.** P. G. Glaskov, A. M. Ofengenden, I. I. Druzhinin, R. P. Nesterovich and G. G. Chepurnoi. (*Stal'*, 1958, (3), 209–213). The Mn content during refining is 0·02–0·04% less than with normal pig-iron though the degree to which the metal can assimilate it is increased from 31·8 to 42%. Desulphurization conditions are less good and S at melt down increases by 0·004%, increasing the refining period and reducing productivity about 1%. S increases with increased S in the iron, with increase of charging period and of heating up time. With ordinary pig iron charging time and content up to 0·05% S have no effect on content at melt down. Low-Mn iron does not effect desulphurization during refining nor P removal nor slag formation, but it increases yield of sound ingots by 0·3% and reduces ore consumption by 7·5 kg/t steel with an increase in Fe–Mn consumption of 1·1 kg/t. Steel quality is unchanged and there is a saving of 11 roubles/t on cost.—R. S.

**Controlled Rimming Steels.** G. de Rousiers. (*Centre Doc. Sidér. Circ.*, 1958, **5**, (2), 377–402). A comprehensive investigation into the control of the rimming action in open and bottle top moulds is presented. Factors influencing the formation of internal defects are considered. Segregation of C, Mn, S, P and N has been studied. The effect of Al additions is discussed. The difference in yield and segregation of metalloids between open top and bottle top ingots is considered.—B. G. B.

**Experience In The Production of Low Alloy Steel in the Zaporozhstal' Works.** I. M. Leikin, I. S. Marakhovskii and A. A. Podgorodetskii. (*Metallurg*, 1956, (4), 13–15). The practice used for the production of a low alloy steel 14KH G'S (1% Mn, 0·6%Si, 0·65 Cr) for the manufacture of welded pipes is outlined.—V. G.

**An Improvement in the Technology of Production of Free Cutting Steel.** I. I. Fomin, A. D. Zaitseva and P. P. Kon'shin. (*Metallurg*, 1957, (1), 15–16). Improvements obtained in the technology of production of the free-cutting steel A12 (0·08–0·16% C; 0·60–0·90% Mn; 0·15–0·35% Si; 0·08–0·20% S; 0·08–0·15% P) smelted in 75-t O.H. furnaces are described. The importance of the following points is stressed: the duration of teeming of an ingot should be not less than 3 min; boiling of the metal should be energetic and the bath should be well deoxidised with ferrosilicon (7–10 kg/t); the ratio of Mn : S should be maintained not lower than 7·5; sulphur should be added to ladle in a metallic container.—V. G.

**Solidification Characteristics of Continuously Cast Low-carbon Steels.** M. Tenenbaum, C. F. Schrader, and L. Mair. (*JISI*, 1959, **191**, Jan., 20–33). [This issue].

**Vacuum Pouring of Ingots for Heavy Forgings.** J. H. Stoll. (*JISI*, 1959, **191**, Jan., 67–80). [This issue].

## PRODUCTION OF FERRO-ALLOYS

**Pittsburgh Coke and Chemical Co. Starts Ferromanganese Production.** (*Iron Steel Eng.*, 1958, **35**, March, 164–166). Work on the adaptation of a conventional blast furnace for the production of Fe–Mn and spiegeleisen is described.—M. D. J. B.

**Titanium Alloys.** K. A. Ottander and M. Grounes. (*Jernkontorets Legeringsmonografier*, 1958, (6), pp. 34). Ferrotitanium and ferro-silico-titanium are amongst the alloys treated. The manufacture of low, medium and high carbon Fe–Ti and of Fe–Si–Ti and their physical properties, chemical composition and application are described. (70 refs.).

**Molybdenum Alloys.** K. A. Ottander and S. Thomas. (*Jernkontorets Legeringsmonografier*, 1958, (7), pp. 33). Amongst other Mo alloys, the methods of manufacture, physical and chemical properties and composition, and applications of ferromolybdenum are described. (50 refs.).—L. D. B.

**Utilisation of Copper Chromium Alloys for the Contact Pieces of Furnaces for the Manufacture of Ferroalloys and Calcium Carbide.** G. Lafeuille. (*J. Four Elect.*, 1957, **62**, (5), 173–179). The use of Cu–Cr alloys for clamps and other electrical contacts in industrial electric furnaces is described, the composition, structure and properties of the material are compared with those of 85/15 brass, and practical details of its use are described.

## FOUNDRY PRACTICE

**Recent Developments in the Manufacture of Castings.** J. L. Rice, R. W. Ruddle and P. A. Russell. (*Proc. I.M.E.*, 1958, **172**, (4), 133–160). Cast-iron and non-ferrous alloys but not steel are reviewed. Established and new processes, grey and malleable iron and nodular cast iron are considered and methods of inspection are outlined.

**Inverse Greyness in Malleable Iron.** G. Östberg and C.-E. Gustafsson. (*Gjuteriet*, 1958, **48**, Jan., 8–11). Investigation of causes of a grey surface rim in white iron castings has been carried out by measurements with thermocouples. The cooling rate in horizontally cast cylinders in the critical temperature range was lower in the surface zone than in the interior. This may account for "inverse greyness." A partial study has been made of other hypotheses.

**Comparison of Liquid- and Air-Quenched Pearlitic Malleable Irons. Part I. Duplex Iron.** AFS Pearlitic Malleable Committee 6-E. (*Trans. A.F.S.*, 1956, **64**, 91–97). Practice of three plants is compared and structures, hardnesses and mechanical properties are measured. Liquid quenching and drawing gives a higher yield strength than does air quenching at a given hardness. (30 references).

**The Use of Steel Melted in an Acid Open-Hearth Furnace for Foundry Purposes.** P. Lwowicz. (*Prz. Odlew*, 1957, **7**, Aug., 219–225). [In Polish]. In Poland, because of the limited amount of pure scrap free from S and P, only steel obtained from an acid O.H. furnace is used in the foundry. The use of this steel is justified by: the possibility of lengthening the pouring period; the decrease in the amount of damage during pouring, the decrease of defects caused by gas blowholes, decrease of repairs, and the cost of the ladle and furnace.

**A Modern Foundry for the Manufacture of Small Steel Castings by New Moulding Techniques.** J. H. Osborn, T. A. Marsden and R. F. Horton. (*Brit. Found.*, 1958, **51**, Feb., 64–76). The effects of the introduction of the shell-mould, Osborn-Shaw and $CO_2$ processes are considered in the layout and construction of a new factory. Heat-treating and control facilities are described.

**Inoculation of Cast Steel.** J. Pakulski. (*Prz. Odlew*, 1957, **7**, April, 97–101). [In Polish]. Two theories of inoculation are

given: the adsorption theory and the theory of making the metal "turbid." The effects of Ca, Mg, B, Ce and $N_2$, on the surface tension is discussed, as well as crystallization and the mechanical properties of various types of steel.

**Graphite Moulds, Pressure Pouring System Casts Steel Freight Car Wheels.** R. E. Ford. (*Western Metalw.*, 1958, **16**, Feb., 46–49). An account of a new wheel foundry of the Griffin Wheel Co., Colton, Calif. The processes used are described, especially the degree of automation.

**Production of Steel Castings with Exothermal Inserts for Head Shrinkage.** I. M. Barabash and A. S. Shuper. (*Lit. Proizv.*, 1957, (12), 4–5). Because of the high percentage of waste at the Novo-Kramatorsk Machine Construction plant, due to feeder head shrinkage, tests were arranged for producing a suitable exothermic mixture which would maintain the metal in the moulds in liquid condition for a longer period. The mixture selected consists of 56·5% charcoal; 7·6% sawdust; 6·22% iron scale, 4·3% ferrosilicon (Si 75%); 2·8% Al powder and 22·6% waterglass. This mixture is first suitably prepared and mixed before addition.—L. H.

**Manufacture of Cast Steel Plough Shares with Alloy Cutting Edges.** I. A. Nilovskii. (*Lit. Proizv.*, 1955, (8), 13–14). [In Russian]. A description is given of the production of ploughshares from Bessemer steel containing C, 0·48; Si, 0·48; Mn, 0·93; S and P, 0·044%.—L. H.

**The Production of Foundry Sands.** A. R. Abraham. (*Brit. Found.*, 1958, **51**, Feb., 59–64). Sources in Britain are surveyed and the characteristics of various deposits are given and the development of blended mixtures considered.

**Calculations for Plants for the Preparation of Moulding Sands.** R. Chudzikiewicz. (*Prz. Odlew.*, 1957, **7**, Oct., 277–281). [In Polish]. Production programmes of foundries and unit consumption of moulding sands are the basis for calculations of plants for the preparation of moulding sands. The consumption data for sands from the best working foundries are given. Classification of plants according to the type of mixers used is given.

**A Review of Equipment Produced in Poland for the Mechanisation of the Conditioning of Moulding Sands.** W. Szanda. (*Prz. Odlew.*, 1957, **7**, May, 135–140). [In Polish]. Different types of sand conditioning plants (which can also be used in old foundries) produced by the Foundry Machines Works in Krakow are discussed and illustrated with diagrams.

**Gas Evolution from Core Sands and Binders.** J. Ornst. (*Prz. Odlew.*, 1957, **7**, Nov., 317–326). [In Polish]. Tests on the volume of gas evolved from core sands show that there is a linear relationship between the binder contents in the mixture and the gas content of the sand. Binders are classified according to the specific volume of gas evolution and the strength properties. The migration of binder to the external layers of the core during drying and its effect on the volume of gas evolution was investigated. The relation between critical volume of gas evolved and critical permeability was determined.

**Jolt Test for Sand.** R. W. Heine, E. H. King and J. S. Schumacher. (*Trans. A.F.S.*, 1956, **64**, 415–421). A test is described and its limitations discussed as well as its relationship to the choice of moulding machine.

**Statistical Techniques for Classifying Foundry Sands.** J. M. Leaman and D. C. Ekey. (*Trans. A.F.S.*, 1956, **64**, 679–687). An ideal system of classification is defined and a method approximating to it is proposed to indicate, by three numbers, grain size and size distribution. Sieve tests and ordinary standard tests are used.

**The Problem of Standards for Testing Moulding Materials.** L. Lewandowski. (*Prz. Odlew.*, 1957, **7**, June, 169–174). [In Polish]. The author describes characteristics concerning Polish departmental standards for laboratory testing of moulding materials and suggests changes. Further, he discusses the methods for grain-size determination and the problem of establishing a uniform nomenclature for permeability ($cm^4/g.$ min).

**Anionic or Cationic Agents: A Solution to Sand Problems.** C. J. Vingas and A. H. Lewis. (*Trans. A.F.S.*, 1956, **64**, 453–458). Variations in surface drying rate of bentonite-bonded moulds were found to be due to changes in moisture migration rate. Addition of surface-active agents improve the sand. Variations using different base sands are attributed to hydrogen ions and this is supported by measured rates of drying of acid-treated sand.

**Use of Glass Spheres for Calibrating Sand Testing Sieves.** R. E. Morey and J. H. Schaum. (*Trans. A.F.S.*, 1956, **64**, 443–446).

**Some Observations on the Transverse Test at Elevated Temperatures for Moulded Sand Mixtures.** D. C. Williams. (*Trans. A.F.S.*, 1956, **64**, 353–356). The transverse test accessory for the Dietert sand strength machine was used. Tests at 1600° and 2000° F were made by timing rupture under the cores' own weights.

**Stickiness in Core Sand Mixtures.** D. S. Mills. (*Trans. A.F.S.*, 1956, **64**, 344–352). Progress Report No. 2 of the 8-C-f Committee. The method of measurement is described and green-strength values are given. Lower moisture cores were less sticky, retained sand correlated with stickiness but geometer tests showed little agreement. The work was applied in the foundry by reducing moisture content with improvement of stickiness, flowability and density. The force required to pull a pattern from the core sand did not correlate with stickiness.

**Bentonite: Properties and Composition. Their Relation to Casting Defects.** W. D. Emmett. (*Trans. A.F.S.*, 1956, **64**, 482–488). Defects produced on repeated re-use of bonded sand are discussed. Various tests on the bentonites are reoprted.

**Koedoespoort's Metal Pattern Shop.** (*Eng. Foundryman*, 1957, **22**, Dec., 38–40). The South African Railways workshops are described.

**Use of Epoxy Resin as Pattern Material.** E. J. McAfee. (*Trans. A.F.S.*, 1956, **64**, 504–508).

**Core-Box Design and the Utilisation of the Mechanical Properties of Wood.** V. Beran. (*Slévárenství*, 1957, **5**, (12), 364–369). [In Czech]. The fullest utilization of the strength of wood in core-box design permits economies in wood consumption. Designs facilitating such economies are considered, and the principles on which their development was based are explained.—P. F.

**Distance Between Feeder-Heads in Castings of Equal Thickness.** M. Notte. (*Fonderia Ital.*, 1957, **6**, Dec., 463–468). [In Italian]. The author approaches the problem of feeder head distribution from a theoretical view-point backed up by practical considerations. A code of practice is given.

**Casting Techniques, Feeders and Risers.** D. F. O'Connor and C. Morgani. (*Fonderia*, 1957, **6**, Nov., 485–490). [In Italian]. The authors discuss casting procedures for iron, casting speeds and temperatures, and the positions and sizes of feeders and risers. A code of practice is recommended.—M. D. J. B.

**Risering of Nodular Irons. III. The Effect of Pouring Temperature on Shrink Depth.** R. C. Shnay and S. L. Gertsman. (*Trans. A.F.S.*, 1956, **64**, 271–283). The shrink depth of hypereutectic nodular castings is greatly affected by pouring temperature. There is a critical temperature above and below which shrink depth increases, this was found to lie between 2325–2375° F. Hypoeutectic irons are much less sensitive. Insulating covers decrease the pouring temperature effect.

**The Intermediate Metal Thermocouple in Metal/Mould Interface Temperature Measurements.** D. V. Atterton and D. H. Houseman. (*Brit. Found.*, 1958, **51**, Feb., 77–82). The development of a method of using the molten metal to form the junction between dissimilar metals is described and 280-lb steel castings were examined by the technique. Calibration showed errors due to contamination to be negligible and initial freezing and subsequent remelting of the metal initially entering the mould was demonstrated. Comparisons with sheathed couples were made.

**Surface Coating for Steel and Iron Castings.** G. I. Tsiganenko. (*Lit. Proizv.*, 1955, (9), 27). [In Russian]. At the Nikolaevski plant the use of a coating for moulds has made possible the production of castings with clean, smooth surfaces without blemishes and without fusion of the sand. The composition is sand 95%, refractory clay 2% and crude oil 3%.—L. H.

**Investigation into the Quality of Castings in Relation to the Heat Resistance of Mould and Core.** N. P. Nikolaichik. (*Lit. Proizv.*, 1955, (9), 20–22). [In Russian]. It was supposed that the low strength of moulds filled with fire brick-clay mixtures, and their cores, was due to the rapid cooling of the fire-brick mixture. Exhaustive enquiries have been made to investigate this matter and these are reported here. They extend from laboratory analysis of moulding materials to large scale investigations under conditions of production. The findings show that to improve the quality of the moulds it is essential to arrange that the crystallization of the castings takes places outwards, and therefore the greater the heat resistance, and the smaller the thermal capacity of the mould, the better the results will be.—L. H.

**Moulding with a Sand Slinger.** G. Lempicki. (*Prz. Odlew.*, 1957, **7**, July, 185–192). [In Polish]. The author describes the technology of sand-slinger moulding based on his own researches. The influence of different factors on the density ratio of moulding sands and the properties of mixtures densified by the slinger are discussed.

**Evaluating the Green Properties of Cereal Core Binders.** E. C. Zuppann and H. Putz. (*Trans. A.F.S.*, 1956, **64**, 267–270). Jolt overhang and jolt sag tests were used. It was found that green compressive strength did not distinguish differences in cereal quality but that these tests are suitable. Core oils have a marked effect.

**Tertal—A Core Binder.** J. Buciewicz and T. Rzepa. (*Prz. Odlew.*, 1957, **7**, Oct., 279–281). [In Polish]. The problem of core binders is discussed. A new core binder Tertal 30/10 and Tertal 20/15 was thoroughly tested and the result of investigations show its high qualities. Technological instructions on how to use the new binder are given.

**The Croning Process; Bibliography.** (*Bücherei und Dokumentationsstelle des Vereins Deutscher Giessereifachleute*, 1956, pp. 16).

**Bibliography of the $CO_2$ Process.** Verein Deutscher Giesserei-Fachleute. (*Rationalisierungs-Kuratorium der Deutschen Wirtschaft-RKW, Frankfurt (Main)*, 1956, (108), pp. 4). Giving 30 references.

**A Survey of the $CO_2$ Process.** J. E. Huss. (*Trans. A.F.S.*, 1956, **64**, 98–99). A brief report from the Core Test Committee is given. The information was derived from a questionnaire from 30 foundries.

**The Hardening of Moulds and Cores with Carbon Dioxide.** A. Bydalek, H. Gumienny and M. Stolarski. (*Prz. Odlew.*, 1957, **7**, June, 161–165). [In Polish]. The method of hardening of moulding sand, containing water-glass as binder with $CO_2$, consists in the formation of hydrated silicic acid. The results of research in this field are given: The rate of hardening increases with the modulus of water-glass. For large moulds, water-glass with a modulus of 2·4–2·8% is recommended, for small moulds a modulus of 3·0–3·5%. The optimal content of water-glass is 4–6%. The density of water-glass and the addition of NaOH increases the strength. The strength is also increased by addition of clay.

**Developments of High Pressure Moulding with $CO_2$ Process Sands.** T. E. Barlow. (*Trans. A.F.S.*, 1956, **64**, 336–338). A progress report on the making and use of production moulds is given.

**Ceramic-Mould Process for Steel Castings.** D. C. Ekey and E. G. Vogel. (*Trans. A.F.S.*, 1956, **64**, 439–442). Procedure at the Lebanon Steel Foundry is described and advantages

and disadvantages are given. A table of comparisons with investment, green sand and shell castings is added.

**Design of Castings with Regard to Pouring.** A. Östberg. (*Gjuteriet*, 1958, **48**, Mar., 33–36). The 4th Report of the Castings Design Committee of the Sveriges Mekanförbund. Reduction of wall thickness by avoiding oblique curves and changes in direction of flow, space for gates and risers, substitution of sloping for horizontal surfaces and improved reinforcement rib design are advocated.

**Heavy Casting at the Newcastle [N.S.W.] Steel Works.** S. G. Urane. (*Castings*, 1958, **4**, Feb., 7, 9, 11, 13, 15). A general illustrated account of the moulding of rolls for the 36 in. structural mill weighing 38 tons cast weight is given.

**Investment Casting: Bibliography.** (*Verein Deutscher Giessereifachleute, Düsseldorf*, 1956, pp. 13). Giving 52 references.

**High Speed Steel (SW18) Cutters Cast Centrifugally in Metal Moulds.** S. Gebalski. (*Prace Inst. Mech.*, 1956, **5**, (17), 12–33). [In Polish]. The method of centrifugally casting cutters is discussed. The technology of the production of ordinary and bimetal cutters by this method was investigated. Data of the chemical and physical properties and durability of the cutters are tabulated. A critical survey of the results obtained is given.

**A Machine for the Centrifugal Casting of Sink Pipes.** H. Mastalerz. (*Prz. Odlew.*, 1957, **7**, May, 140–144). [In Polish]. The methods and economic basis for the production of sink pipes centrifugally cast in sand moulds are discussed. A description with diagrams of the machines in use and their operating conditions are given.

**Concerning Shell Moulding.** L. Petržela. (*Slévárenství*, 1957, **5**, (12), 99–104). [In Czech]. Researches on the effects of design, composition and binder quality on the strength of shell moulds and the surface quality of ferrous castings are described, and some generalizations based on the results obtained are discussed.—P. F.

**A New Process of Shell Moulding.** C. Stefanescu and I. Gentiu. (*Met. Constr. Masini*, 1957, **9**, (8), 1–8. [In Rumanian]. The composition of the sand-resin mixture is described in detail. Two qualities of native sand have been investigated at the Institute of Machine Construction, and the moulding mixture forced in by compressed air with an addition of waterglass. A very detailed study is added and the introduction of this method is recommended.—L. H.

**More Foundries Discover Shell Moulding Can be Simple.** (*Can. Metalw.*, 1958, **21**, Feb., 30–31). A brief detailed account of the process as operated by the Toronto Foundry Co.

**The Principles of Pattern Plate Design for Shell Moulding.** H. Rapacki. (*Prz. Odlew.*, 1957, **7**, Nov., 309–317). [In Polish]. Working conditions of pattern plates for shell moulding and the materials used for the production of components of these plates are described. The principles of designing pattern plates, accuracy of dimensions and surface smoothness of the working models, their thermal capacity and resistance to distortion are discussed. Examples of typical pattern plates are given. Different designs of ejector pins and sets of ejectors of fixing elements are shown and discussed.

**The General Motors Blow-Hot Press Automatic Shell Moulding Machine.** R. S. Amala, J. H. Smith, A. L. Boegehold and R. F. Thomson. (*Trans. A.F.S.*, 1956, **64**, 422–429). A method using blowing techniques for large uniformly contoured shell moulds and a hot pressing process for curing is described and also the construction of machines for carrying out the cycle. Moulds for automotive camshafts are produced.

**Analysis of Factors Affecting Surface Finish of Grey Iron Castings.** R. L. Yard and D. C. Ekey. (*Trans. A.F.S.*, 1956, **64**, 671–678). Sand fineness, static metal pressure and mould hardness were studied. It was found that sand fineness and metal pressure interact to influence surface finish, roughness increases with decrease of the former and increase of the latter. Mould hardness due to ramming does not influence surface finish significantly, nor does sand fineness in comparison with interaction. Suggestions for further studies are made.

**Gases in Cast Iron.** (*Bradley's Mag.*, 1958, Spring, 3–4). $O_2$, $H_2$ and $N_2$ are discussed, and their effects briefly considered.

## VACUUM METALLURGY

**Effect of Degree of Vacuum During Annealing on Properties of Transformer Steel.** N. F. Dubrov and V. A. Koroleva. (*Metall. Obrabotka Met.*, 1957, (4), 51). Tests on seven heats of hot-rolled transformer steel 0·019 in. thick showed that when the degree of vacuum used for annealing was reduced from 30–40 mm to 0·005 mm the properties of the steel were considerably improved.—R. S.

**Behaviour in Iron and Steel Melting in a High Vacuum.** W. A. Fischer and A. Hoffmann. (*Forsch. Wirts. Nordrhein-Westfalen*, No. 586, 1958, pp. 41; *Arch. Eisenh.*, 1958, **29**, June, 339–349). The behaviour of $N_2$, $H_2$, P, As, S, O, Mn, Si, Cu, Mg, Ca, Zr and Al in 0·001–3·5% C iron melted in various refractory vessels for 6–7 h and in some cases 20–30 h in 4–5 kg quantities was examined. The inclusions were examined chemically and metallographically and the effects of the refractory materials and others on deoxidation and other reactions were estimated.

## REHEATING FURNACES AND SOAKING PITS

**Where Faster Heating Saves.** C. A. McFadden. (*Steel*, 1958, **143**, Sept. 1, 68–70, 72, 73). The use of accelerated rates of heating for forging, rolling and hardening is briefly considered. Possible savings are pointed out.

**The Reheating of Special Steel Billets.** R. Courdille. (*Centre Doc. Sider. Circ.*, 1958, **15**, (5), 1059–1069). A billet reheating furnace is described which has a conventional pusher-type feed in the heating zone and a rocking hearth in the soak zone. It is claimed that this design of furnace gives much more even temperature distribution through the billet.—B. G. B.

**The Firing of Small Forging Furnaces with Heavy Fuel Oil.** H. Remus. (*Stahl Eisen*, 1958, **78**, July 24, 1038–1041). The experience gained in the change-over in the firing of small forging furnaces from producer gas to heavy fuel oil is described and heat balances are produced. Burners with compressed-air atomization have proved superior to low-pressure burners. The oil consumption falls asymptotically with throughput $m^3$ h.—T. G.

**Heating Without Scale Formation by Means of Lithium Carbonate in a Small Forging Furnace Fired by Grid Gas.** F. Neuberger, L. Rötz and L. Möckel. (*Werkstattstechn. Maschinenbau*, 1958, **48**, June, 341–343). Trials are described in which scaling is prevented by heating at temperatures up to 1200° C in a lithium salt atmosphere. Operational details and the surface properties of the heat-treated forgings are discussed. High alloy steels were treated successfully.

## HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

**Control of Furnace Atmospheres for the Thermal Treatment of Metals.** A. L. Forrester. (*Chal. Ind.*, 1958, **39**, June, 159–166). This is a general survey, which deals with the objects of furnace atmosphere control, the gases used and their properties, and furnace design from the point of view of atmospheric control.

**Gaseous Cementation Using Solid Carburisers.** P. I. Shashkin. (*Metallov. Obra. Met.*, 1955, (6), 40–44). The relative positioning of the articles and the solid carburizers in appropriate boxes is an essential part of the process. It ensures an even saturation of the surface layer throughout its thickness, which is characterized by a uniform hardness.

The process allows a shortening by a factor of 2½ to 3 of the heating time to the temperature required.—L. H.

**Carburizing-Martempering Treatment for Crankshafts in New Type of Salt Bath Line.** (*Ind. Heat.*, 1958, **25**, Feb., 284, 286). An installation consisting of a 360 KW electrode-type carburizing both 178 × 32 × 30 in. an 80 kW neutral salt bath 12 × 30 × 32 in. and a 30 KW draw furnace 52 × 32 × 30 in. wash and rinse tank and push-pull mechanism is briefly described.

**The Rate of Cementation when Heating with High-Frequency Current.** A. D. Assonov, K. Z. Shepelyakovskii and P. A. Lankin. (*Metallov. Obra. Met.*, 1955, (3), 39–50). A comparison is drawn between the usual process of cementation of 8–10 h for gaseous and of 12–15 h for solid carburizers, and the high-frequency heating process which only takes 45–60 min.

A few types of steel have been investigated and results are set out in detail.—L. H.

**Resistance to Cracking of Spheroidal Graphite Cast Iron When Subjected to Electric Hardening.** (*Metallov. Obrab. Met.*, 1956, (12), 56–60). With low phosphorus content (0·1% max.), surface hardening of pearlitic and ferritic cast irons can be effected at all quenching temperatures. With P at 0·17–0·20%, the surface hardening temperature must not exceed 1000° C for the pearlitic and pearlitic-ferritic irons while the ferritic and ferritic-pearlitic irons can be surface hardened at 1000° C and higher. High P (0·3%) iron is not recommended for parts which have to be surface hardened. The tendency to cracking increases with increasing amounts of pearlite in the initial structure and also with hardening temperature.—R. S.

**Russian Induction Hardening and Forming.** K. Z. Shepelyakovski. (*Metalw. Prod.*, 1958, **102**, Mar. 21, 501–504). Automobile gear box treatment is described. High-frequency heating is used.

**Conditions to Be Avoided in the Induction Heating of Steel for Hardening.** I. N. Kidin. (*Metallov. Obrab. Met.*, 1956, (5), 25–27). The decisive factor which causes variations in the heating curve reflecting the kinetics of induction heating is the magnetic transformation of ferrite. In induction heating, when the transition of ferrite from the ferromagnetic to the paramagnetic state occurs, the changes which take place in the electromagnetic system involving the inductor and the object which is being treated cause a marked retardation of the increase in temperature of the surface layer. The relationships between the formation of austenite and the subsequent heterogeneity of martensite after quenching, on the one hand, and the kinetics of induction heating on the other hand, are described. In the application of high frequency hardening it is not possible to avoid discrepancies in the results if it is carried out from the temperature of the isothermal stage. The temperature range to be avoided in order to ensure stability of results is shown graphically.—R. S.

**Low Temperature Nitriding of Structural Steels.** R. I. Mishkevich, P. I. Solntsev and A. V. Smirnov. (*Metallov. Obra. Met.*, 1957, (2), 49–54). A process operating below 400° C to avoid temper-brittleness in the core has been developed. Ferrosilicon is added to promote $NH_3$ dissociation, times up to 80 h are used to obtain thick nitrided layers. Hardnesses and penetrations are shown. $CrO_3$ is used to raise surface temperature.

**Effects of Nitriding on the Erosion Properties of Low Alloy Steels for [Working] Components of Steam Power Equipment.** A. D. Moiseev. (*Vestnik Mashin.*, 1954, (4), 61–65). Experimental data show that, nitriding cannot serve as a reliable means for increasing resistance to erosion of low alloy steels against the impact of jets or currents of the working fluid passing at great speed through narrow gaps. In the absence of such conditions, the nitriding of these components is beyond doubt effective. In such cases the positive results of nitrided surfaces in their resistance to friction are well established.

**The Re-Nitriding of Steel.** G. J. Cox. (*Met. Treatment*, 1958, **25**, Feb., 47–51, 68; Mar., 101–106). Trials on three steels are described, Cr–Al–Mo, Cr–Mo and Cr–Mo–V

were denitrided and renitrided and the changes in properties determined. Structures are shown.

**Gaseous Carbo-Nitriding of Rings for Spinning and Twisting Machines.** V. N. Saikina and M. P. Kalyanova. (*Vestnik Mashin.*, 1957, (10), 69–70). [In Russian]. For two years comparative experiments have been carried out with rings subjected to gaseous carbo-nitriding and to nitriding with solid carburizers. The findings proved that rings with gaseous carbo-nitrided showed a fourfold increase in wear resistance in cotton spinning machines, and 1·5 times in silk spinning machines.—L. H.

**Decarburisation of the Surface of Castings.** I. V. Trush. (*Lit. Proizv.*, 1955, (9), 27–28). [In Russian]. Analysis has shown that objects from steels 40 and 40 KH obtained by investment casting, with carburizers, do not have a decarburized surface in contrast with castings obtained by the usual methods.—L. H.

**Decarburisation Annealing of Cast Iron Sheet.** E. G. Shumovskii. (*Metallov. Obrab. Met.*, 1956, (7), 10–15). The continuous casting and rolling of cast iron to produce 0·8 to 1·0 mm sheet has been developed in the U.S.S.R. The sheets have the structure of white cast iron and are, therefore, subjected to a graphitizing annealing. To obtain properties equal to those of conventional roofing tin, this continuously produced iron sheet must be fully decarburized. The carbon in the surface layers is oxidized whereupon the carbon of the core diffuses to the surface, is oxidized and removed, passing into the furnace atmosphere. This article deals with the problem of improving white iron sheets containing initially 3·2 to 3·5% C, by transformation to purely ferritic mild steel containing 0·08 to 0·05% C with improved properties similar to those of cast iron sheet. Control of furnace atmosphere is essential.

**Large Oven Furnace Program—Controlled for Versatile Heat Treating.** J. H. Helfat. (*Ind. Heat.*, 1957, **24**, Dec., 2512, 2514, 2516). A gas-fired installation operating from 600–2000° F for stress-relief and similar duties with internal dimensions 6 × 6 × 6 ft is described.

**Heat-Treating an Outsize Gear Broach.** (*Metalw. Prod.*, 1958, **102**, Mar. 14, 457). A 1025 lb forging 69 ft long and 9⅛ in dia was treated. Few details are given.

**Annealing Malleable Cast Iron in Liquid Media.** W. Sakwa and J. J. Jackowski (*Prz. Odlew.*, 1957, **7**, April, 93–96). [In Polish]. Preliminary investigations on the annealing of pearlitic and ferritic malleable cast iron in $BaCl_2$ and NaCl were conducted in order to verify the results obtained in the U.S.A. and Russia. Pearlitic malleable cast iron obtained after 11 hours of annealing had tensile strength $>$ 40 kg/mm² and elongation $A_2 > 5\%$ and ferritic malleable cast iron after 16 hours shows $>$ 30 kg/mm² and elongation $A_2 > 11\%$. Further investigations are recommended.

**Mechanical Properties and Microstructure of Steels 25 L and 20 L in Relation to Their Thermal Treatment.** E. P. Ivanyushin and K. N. Kulikova. (*Lit. Proizv.*, 1955, 11), 1–4). [In Russian]. The investigation extends to various thermal treatments such as the temperature and duration of annealing, quenching in water, normalization and subsequent tempering for removing stresses, etc.—L. H.

**Stress Relief of Austenite Stainless Steel Strips.** F. de Kazinczy. (*Jernkont. Ann.*, 1958, **142**, (1), 5–14). Relief of residual stresses in tempering was studied by determining internal stress distribution. Body stresses decreased with the increase in temperature from below 100° C, but those produced by balancing external forces were only reduced at much higher temperature. For convenience a body stress can be considered as the mean of a textural stress in a thin layer. Further decrease of stress at 300–400° C is probably due to creep.

**Stress-Relieving Welded Teeming Ladles. Combined Weight with Metal Over 400 Tons.** J. Finn. (*Iron Coal Trades Rev.*, 1958, **176**, Feb. 21, 457). The construction of the ladles at Port Kembla is described.

**Rational Conditions for Annealing Rolled Sections in Batch Furnaces with Mechanised Charging in Packets.** M. I. Kolosov, I. Ya. Aizenshtok and A. I. Komissarov. (*Stal'*, 1958, (1), 71–74). Statistics are quoted for the increase in annealed metal during 1946–1956 with reduced fuel and power consumption and in cost per t. The Chelyabinsk and Dneprospetsstal plants are in advance of the movable-hearth furnaces at other works.—R. S.

**Investigating Isothermal Annealing of Some Alloy Steels.** E. S. Tovpenets (*Metallov. Obrab. Met.*, 1956, (2), 53–56). The stability of supercooled austenite depends not only on the chemical composition of the steel but on a number of factors, chief of which are the heating temperature, time of holding and rate of cooling High austenitizing temperature improves the stability of supercooled austenite. Step annealing is not satisfactory for a number of alloy steels. The period of holding in step annealing must be increased more than is indicated by the isothermal breakdown curves. Conditions for completing the breakdown of austenite are described.

**Efficient and Economic Conditions for Annealing Blanks of Steel 5 Kh NT.** N. A. Timchenko. (*Metallov. Obra. Met.*, 1955, (4), 32–38). In investigations arranged to observe the kinetics of the transformation of austenite in the type 5 KhN T steel, it was found that the optimum temperatures are, for the first stage, 675–650° and for the second, 375–275°. Within these ranges the breakdown of austenite is nearly complete (90%) inside one h. Suggestions for the annealing of this steel which are proposed, are the result of extensive tests, results of which are detailed in the article.

**The Annealing of Cold Rolled Rimming Steel Sheet Containing 0·08 to 0·10% Carbon and 0·30 to 0·40% Manganese.** V. K. Barzii and S. S. Kolot. (*Stal'*, 1958, (2), 159–161). In the annealing of cold rolled rimming steel sheet, an increase in the ferrite grain size is observed when the annealing temperature (up to the temperature of phase transformation) does not cause any further increase in the ferrite grain size. The yield point is the most sensitive mechanical property in relation to annealing conditions.—R. S.

**Study on High Speed Tools. XIX. Stepped Annealing of High Speed Steel SKH6 and SKH8.** H. Hotta. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1071–1073).—K. E. J.

**The Effect of Bismuth and Boron on the Reduction of the Time for Annealing White Cast Iron.** I. F. Kurtov, V. A. Sakharov, N. P. Chichagova and S. V. Ryabokon'. (*Lit. Proizv.*, 1957, (12), 20–21). The joint addition of small quantities of bismuth and boron makes it possible to shorten the duration of annealing by 25–30%, it doubles the limits of chromium content in the white iron without increasing the duration of annealing, and simplifies the technology of the production of malleable iron with pearlite grain structure.—L. H.

**Variables Affecting Continuous Strip Annealing Temperatures Constantly Balanced by Precise Temperature Control System.** (*Ind. Heat.*, 1958, **25**, Feb., 314–316, 318). The installation of the Steel Co. of Canada, Hamilton, Ontario, is described and the control systems shown in a diagram and briefly described.

**Improving the Quality of Mild Steel Sheet By a Very Simple Heat Treatment.** V. A. Konopasevich. (*Metallov. Obrab. Met.*, 1956, (3), 55–59). Simple heat treatment tests of MSt 3 (0·14–0·22% C) rimming steel showed that normalizing and blowing the sheets with air improves their impact strength at temperatures above and below zero. Still better is the water-spray cooling of sheets heated to 900° C. The microstructure becomes fine-grained and there is a much greater increase in impact strength than in normalized sheets. Tempering is not essential after hardening. Experiments are being carried out to ascertain whether the sheets can be water-sprayed on the roller table immediately after rolling.—R. S.

**Oils for the Quenching of Steel** H. E. Priston. (*Met. Treatment*, 1958, **25**, Feb., 57–64). A review of quenching and quenching fluids. The three stages (vapour blanket, vapour transport and convection cooling) are distinguished and the evaluation, and handling of oils are outlined as well as the ordinary heat-treating processes.

**A New Device for Evaluating Quenches.** R. M. Critchfield. (*Steel Proc.*, 1958, **44**, Jan., 42). A brief account of the magnetic Quenchometer is given. A heated ball falling into a quenching liquid is attracted to a magnet on reaching the Curie point and a timing device is incorporated.

## FORGING, STAMPING, DRAWING AND PRESSING

**Allegheny Ludlum Steel Corp.'s New Forge Shop, Dunkirk, N.Y.** E. Kern and J. E. Trabits. (*Iron Steel Eng.*, 1958, **35**, June, 73–80). The authors describe the new 2000-t general purpose forging press recently installed and the ancillary plant and services.

**A Modern Forging Press and its Control.** O. Hoffmann and A. Krynytzky. (*Iron Steel Eng.*, 1958, **35**, June, 81–93). The characteristics of a modern forging press are reviewed and a detailed description is given of a 2000-t pull-down, two-column press. Drive, press controls and performance are discussed.

**Metal Flow in Forging. VI. High Temperature Compression Tests of Metals.** T. Awano and M. Ochiai. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, Sept., 469–472). High-temperature low-speed tests on several kinds of steel and other metals were carried out at rates of $10^{-3}$–$10^{-2}$sec. Maximum real stress was observed at 10–20% compression ratio and was increased by higher rates and lower temperatures. Carbon content has no effect above 1000° C. A small increase appeared at austenizing temperature. A power law is reasonably accurate for real stresses, which also decrease slightly on long-time heating or normalizing.

**Power Requirements in Drop Forging.** H. Lippmann and F. Neuberger. (*Werkstatttechn. Maschinenbau*, 1958, **48**, Aug., 449–452). A recently published method is tested for calculating the power required in drop forging. It is shown theoretically and by comparative examples that a minimum value can be furnished, which can be exceeded substantially.

**Shape and Positional Tolerances. II.** J. Ickert. (*Werkstattstechn. Maschinenbau*, 1958, **48**, Aug., 428–434). Individual examples of deviations from shape (i.e. from the circular form) and of position (i.e. deviation of a line or surface from the desired position) are given, together with methods of measurement. The results are discussed.

**Extra Forging Steps Yield Special Benefits.** H. Chase. (*Iron Age*, 1958, **181**, May 1, 81, 83). Examples are given in which special processing at the forging stage has allowed savings to be made in subsequent treatment.—D. L. C. P.

**The Life of Forging Dies.** J. Elfmark. (*Hutn. Listy*, 1958, **13**, (7), 612–619). A detailed experimental and statistical study of the principal causes of die wear in the case of various classes of forging dies is described; the heat-treatment of the die steels, their chemical composition, the manner of using the dies in forging, their cleaning, lubrication, cooling, shape, and design tolerances were considered. Dies wear most at outer edges and can therefore be repaired by local weld-surfacing with alloy electrodes.—P. F.

**Low-Cost Conveyor Units Speed Forging Lines.** R. H. Eshelman. (*Iron Age*, 1958, **182**, July 10, 110–112). A system of conveyors developed by a Jackson, Mich. Company for a presswork line is described.

**Vibration Studies in the Vicinity of Forging Shops.** H. W. Koch. (*Stahl Eisen*, 1958, **78**, Sept. 18, 1320–1326). The amplitudes of vibrations in the vicinity of forging shops were measured. The recording methods are described and the theoretical principles discussed. The effects of the vibrations on buildings and human beings are explained.

Characteristic results for various types of hammers are discussed and distances are quoted at which the vibrations are sufficiently reduced.—T. G.

**Forging and Stamping of Heat-Resisting Alloys.** D. I. Berezhkovskiĭ. (*Metallov. Obrabotka Met.*, 1957, (8), 51–61). This is a survey based on 18 British, American, German and Belgian references, of methods and problems of forging and stamping of heat resisting alloys. Metallurgical problems are referred to only as they influence the behaviour of the ingot in forging.—S. K.

**One-Piece Wheels in Special Steel for Railway Equipment.** (*Aciers Fins. Spéc.*, 1958, July, 42–46). The casting, forging, heat treatment and machining are described of wheels for autorail cars, manufactured from Cr–Mo steel.

**Increasing the End-of-Stamping Temperatures as a Means of Lowering Stamping Force for Large Parts.** A. V. Altykis and N. I. Belan. (*Metallov. Obrabotka Met.*, 1957, (10), 48–52). It has been shown that by increasing the temperature at the end of stamping from 850 to 950–1050° C the required stamping force can be reduced by $\frac{1}{4}$ or $\frac{1}{2}$ for type 40Kh, 40N, 43N, 34KhM or 34KhN3M and other steels. This is based on measurements of ultimate strengths of the steels at 825°, 925° and 1025° C. Microstructural investigations have shown that heat treatment conditions for correcting overheating of large forgings can be formulated which enables the higher stamping temperatures to be used safely. The higher temperatures, combined with better lubrication and other measures have enabled stamping to replace forging in many applications.—S. K.

**The Hot Deformation of Cast Iron.** H. Meyer. (*Werkstattstechn. Maschinenbau*, 1958, **48,** June, 346–348). The conditions necessary for satisfactory hot pressing of cast iron are outlined. These include the type of graphite, maximum permissible forging temperatures—influenced by phosphorus content—nature of the forging process and design of the press.

**Improving the Quality of Steel Used for Cold-Pressing of Bolts.** D. Tlustá. (*Výzkumné Práce z Oboru Oceli a Zeleza, State Publ. Ho. Tech. Lit., Prague*, 1958, 173–188). [In Czech). Medium-carbon steels containing about 0·2% Cu were produced from O.H. melts into which oxygen had been blown for about 10 min. towards the end of the period of boil. As a result of the increased homogeneity of the steel, resulting from the use of $O_2$, and primarily due to the reduction in the amount of inclusions and in the gas-content, a steel of greater ductility was obtained. The added ductility helped to reduce the reject rate in pressing of bolts. Surface quality was improved by modifications in the rod-rolling procedure.—F. F.

**Determination of the Method of Optimum Division of a 8·5 t Ingot and Classification of Slabs for the Deep-drawing Steel Sheets.** W. Leskiewicz, R. O'Donnel, S. Spiewak and E. Wosiek. (*Hutnik*, 1957, **24,** Dec., 477–488). [In Polish]. A general characteristic of the deep drawing of steel sheets is given. The principles of division and classification of slabs are outlined.

**The Forming of Flanged and Dished Heads.** H. S. Beers. (*Iron Steel Eng.*, 1958, **35,** May, 110–113). This paper describes the manufacture of flanged heads by spin-flanging on special machines. The range of these presses is from $\frac{3}{16}$ in. to 6 ins or more.—M. D. J. B.

**Steel Shot Blasts Panel Contour.** (*Steel*, 1958, **143,** July 7, 68–70). The bending of large parts such as aeroplane wing panels by peenforming is described. Steel shot driven against the metal causes it to curve away, and leads to better fatigue strength than other bending methods. In the new "Peenamatic" machines a carriage fires the shot as it travels across the wing panel.—D. L. C. P.

## ROLLING-MILL PRACTICE

**The Rate of Deformation in Hot Rolling.** Z. Wusatowski. (*Neue Hütte*, 1958, **3,** Aug., 494–497). A relationship is derived for the rate of deformation during hot rolling, and is illustrated by a numerical example.

**Roll Pass Design for Combination Structural and Wide Flange Beam Mill.** H. E. Muller. (*Iron Steel Eng.*, 1957, **34,** Dec., 85–95). Various roll pass designs are discussed and the best procedures for use on a number of mills are described. Beam section passes for 40 in. and 44 in. universal mills are given and also for 28 in., 32 in. and 34 in. non-universal type mills.—M. D. J. B.

**A Substantial Acceleration of Rolling with Minus Tolerances.** B. B. Rozin and R. S. Geifman. (*Stal'*, 1958, (1), 51–55). The calculation of theoretical weights of rolled products is explained.—R. S.

**Rolling According to Theoretical Weight at the Petrov Plant.** I. I. Entin and N. I. Beda. (*Stal'*, 1958, (1), 55–57). The possibility of saving metal by rolling to minus tolerances is discussed. Tolerances for heavy plates should be extended downwards without lowering mechanical properties and mills should be equipped to record thickness during rolling. A bonus system to stimulate this method of rolling is recommended.—R. S.

**New Design of Roller Guides.** M. M. Shternov and N. F. Gritsuk. (*Metallurg*, 1958, (1), 23–28). The article describes a new design of roller guide for a 12 in. cross country mill for rolling small channels, angles, rounds, squares and flats. The mill consists of ten 2-high stands and rolls at a speed of 9 to 11 m/sec in the finishing stand.—R. S.

**Preparation of Chromium-Plated Plain Guides for Rolling Mills.** M. A. Tylkin. (*Metallurg*, 1958, (2), 27–28). The article described Cr-plated plain guides which are finding increasingly wide application in Russian mills. They require less maintenance than roller guides and impart a good surface to the stock. Their average life is twenty shifts. The plating and heat treatment of such guides are described.—R. S.

**The Elimination of Sticking of Low Carbon Steel Sheets by the Addition of Chromium.** G. M. Katsnel'son and M. N. Kravets. (*Stal'*, 1958, (2), 156–158). The addition of Cr to low-C steel substantially reduces the amount of rejects in thin sheets caused by sticking during pack rolling. It increases mill productivity, lowers the cost of sheets and eases the work of the rollers. When Cr is added in amounts up to 0·30%, the Mn content must be reduced in order to maintain a normal boil. FeCr is added in the ladle. An increase in the P content of the metal up to the top limit of the relevant Russian specification reduces the tendency of thin sheets to stick when being pack rolled.—R. S.

**Increased Rolling Mill Capacity at Kwinana and Newcastle.** G. D. Henderson. (*Proc. Australasian I.M.M.*, 1957, Dec., 73–98). The whole layout is shown and described, and the electrical equipment, furnaces, coiling equipment and capacity are reviewed.

**Control of Strip Width in Hot Rolling Mills.** (*Engineering*, 1958, **185,** Mar. 21, 382). A note on the B.I.S.R.A. system installed by the Steel Co. of Wales.

**Determination of Spread in Rolling.** M. V. Shuralev and S. G. Nekrasov. (**Stal'*, 1958, (1), 57–60). Experimental and calculated data on spread during rolling in grooves are analysed and the need for trial correction coefficients in recent formulæ is shown. With these corrections the spread in oval and square passes and in smooth rolls is obtained with precision.—R. S.

**Ward Leonard Control of Reel Drives on High Speed Cold Strip Mills.** H. K. Messerle. (*J. Inst. Eng. Australia*, 1957, **29,** Dec., 301–309). Performance and transient characteristics are analysed and sources of instability are identified.

**Production of Thin Cold Rolled Wire.** I. S. Pobedin, V. I. Bairakov, M. G. Uglov and V. G. Drozd. (*Metallurg*, 1958, (2), 32–34). 5 to 6 mm diameter wire is hot rolled and smaller wire is drawn. Due to the well known disadvantages of drawing, research has been done on the cold rolling of thin wire. A twelve-stand continuous fine wire mill is described. Three types of pass design were tried: (1) flat to flat; (2) diamond to diamond; (3) diamond to square. Transitional passes were well radiused to obtain a round section. The second and third types of rolling produced a 1·8 mm diameter wire from 8 mm diameter rods in three passes through the mill without intermediate annealing. Optimum results in rolling 1·8 mm diameter CrNi wire were obtained with two such continuous trains. Roll diameter of the second train was 4 in. (slightly less than in the first train). Good results were obtained and experimental rolling is being carried out to endeavour to increase rolling speed.—R. S.

**The Effect of Rolling Unstable Austenitic 0·76% Carbon Steel at 220–300° C.** P. R. V. Evans and H. O'Neill. (*JISI*, 1959, **191,** Jan., 34–44). [This issue].

**The Effect of Cold Rolling and Heat Treatment on the Properties of Cold-rolled Stainless Strip and its Structure. II.** K. Fukuta. (*Tetsu-to-Hagane*, 1957, **43,** Sept., 1076–1077). The effects of quenching and tempering temp. on the hardness of high- and medium-carbon 13-Cr steels are examined.—E. E. J.

**Production of Hot Rolled Tubes from EI 595 Steel and the Properties of Such Tubes.** N. S. Alferova, G. P. Pishchikov and V. P. Konovalov. (**Stal'*, 1958, (1), 60–66). Details are given of properties during hot working and of the effect of hot rolling temperature and heat treatment on cold brittleness.—R. S.

**Textured Steel.** F. R. Park. (*Prod. Eng.*, 1958, **29,** Mar. 17, 73–78). Roll-embossed steel sheet, its production, finishing and uses are reviewed.

**Percussion Drill Steel Life.** (*Can. Mines, Research and Special Projects Report*, 1957, July-Dec., 17–18). Spiral rolling is found to improve durability.

**Automatic Gauge Control in Rolling Mills.** R. B. Sims. (*J. Inst. Metals*, 1958, **86,** Mar., 289–302). Non-ferrous strip is considered.

## MACHINERY AND SERVICES FOR IRON AND STEEL PLANT

**Planning of Works Transport in the Metal Industry.** J. Polášek and E. Zelinka. (*Hutn. Listy*, 1958, **13,** (7), 626–631). Principles and methods of planning transport systems in large metallurgical enterprises are considered.

**Scrap Processing Plant Makes Use of New Technological Improvements.** (*Iron Steel Eng.*, 1958, **35,** April, 144–149). A description is given of a high-duty scrap baling press, which produces a 60 in. × 20 in. × 24 in. bundle weighing about one ton.—M. D. J. B.

**Appropriate Position and Design of Cabs and Control Gear in Cranes.** F. E. Hänsel. (*Stahl Eisen*, 1958, **78,** Sept. 18, 1326–1332). A detailed account with many graphs and illustrations on the position of cabs, on the design and interior equipment and on the arrangement of the control gear of cranes is presented.—T. G.

**Dual Bridge Drives for Overhead Travelling Cranes.** W. J. Tunny. (*Iron Steel Eng.*, 1958, **35,** April, 125–127). The author describes a number of E.O.T. cranes with individual drives and suggests that, provided the crane is suitably designed, this arrangement has definite advantages particularly for heavy cranes.—M. D. J. B.

**Proper Installation and Maintenance of Electric Control Systems.** J. W. Bauer, jun. (*Iron Steel Eng.*, 1958, **35,** April, 128–131). The article emphasizes the need for careful, well thought out design in electric control systems. The need for good planned maintenance is also stressed.—M. D. J. B.

**Trends in Steel Mill Power Plants.** H. G. Kitt and R. W. Worley. (*Iron Steel Eng.*, 1958, **35,** March, 90–93). The authors discuss power requirements in steel works and in the rolling mill departments especially. They believe that steam conditions of the order of 1250 lb/in² and 975° F may become fairly common in the larger mills.—M. D. J. B.

## LUBRICATION

**Maintenance of Oil Film Bearings on European Mills.** S. Carson. (*Iron Steel Eng.*, 1958, **35,** April, 90–95). Innovations recently introduced in European mills are discussed. These have resulted in improved conditions in bearing shops and lubricating systems.

**Development and Application of Spray Lubrication.** E. J. Gesdorf. (*Iron Steel Eng.* 1958, **35**, May, 115–124). Spray lubrication is an economical method of applying lubricant to open glazing, cables and chains. The paper discusses the design of nozzles, spray-pattern, types of lubricant and systems of application.—M. D. J. B.

**Lubrication and Sulfinuz.** Y. de Villemeur (*Mét. Corr. Ind.*, 1958, **33**, Apr., 171–183). The properties of sulfinuz treated metals and its uses as a lubricant with or without the admixture of liquid or solid lubricants are discussed.

## WELDING AND FLAME-CUTTING

**The Useful Application of the Iron Carbon Diagram in Welding Technique.** F.-C. Althof and G. Hoffmann. (*Schweisstechn.*, 1958, **12**, July, 93–100). This section commences with an outline of the principles underlying the construction of the iron–carbon diagram, followed by a brief account of metallic crystallization and structure and the effect of temperature, time and small alloy additions on the formation of iron–carbon alloys.

**Welding Joints in Steam Pipes: Reliability of Welds Made by the New E.B. Process.** N. P. Pienaar. (*Eng. Found.*, 1958, **22**, Jan., 47–52). An argon-arc root weld insert process is described. A wire is used for insertion and the compositions available for various purposes are given. The procedure and the structures formed are carefully explained and the limitations of the process noted.

**Arc Welding of Structural Grade Steel at Low Atmospheric Temperature.** (*Can. Mines, Research and Special Projects Report*, 1957, July-Dec., 6–9). Welds made between −62° and +21° C were examined for weld-metal cracking.

**The Inert Gas Tungsten Welding of Thin Sheet Steel for Transformers and Dynamos.** J. Ruge and W. Rauchfuss. (*Schweissen Schneiden*, 1958, **10**, Feb., 55–58). Investigations on automatic inert gas tungsten welding of dynamo and transformer sheet containing 0·7–4·3% Si, 0·35 and 0·5 mm in thickness are reported. It is shown that a heat treatment at 930° C after welding will improve properties so as to approximate those of the base metal and conform to DIN 46400 for hot-rolled dynamo and transformer sheet.

**Inert-Gas Tungsten-Arc Welding. SAE 4130 Steel Sheet. Mechanism of Hot Cracking.** C. A. Terry and W. T. Tyler. (*Weld. Met. Fab.*, 1958, **26**, Mar., 103–110). The phenomenon is discussed in detail and the factors involved are set out. The mechanism is considered and factors controlling extent of cracking are suggested.

**Welding an Outsize Stainless Steel Vessel.** (*Metalw. Prod.*, 1958, **102**, Mar. 21, 500). A vessel of 18–8–1 austenitic steel was welded with Fusarc automatic machines.

**Manufacture and Metallurgy of Flash-Welded Line Pipe.** M. A. Scheil, G. E. Fratcher, S. L. Henry, and E. H. Uecker. (*Eng. J.*, 1958, **41**, Feb., 60–71). An illustrated account of production, transport, bending and welding of 16–30 in. dia. pipes, with mechanical test results.

**Construction of Continuous Tube Welding Plants.** Yu. A. Mednikov. (*Metallurg*, 1958, (1), 32–34). The article states the relationship between the first roughing formers and the remaining formers and describes the cooling and dressing devices.—R. S.

**Lime-Coated Electrodes for Welds of High Quality.** A. Schmidt. (*Schweisstechn.*, 1957, **11**, Dec., 143–150). The welding of structural steel for large structures and vessels with lime-coated electrodes is discussed. Recommendations are given for correct electrode positions at different angles. Applications are discussed.—U. E.

**The Different Methods of Examining Welds.** C. Mathy. (*Prat. Sond.*, 1957, **11**, June-July, 102–105). [In French]. The author reviews the methods of testing welds and lists tensile tests transversal to and longitudinal to the weld, bending, resilience and fatigue tests.

**Definition of a First Quality Spot-Weld.** International Institute of Welding, Commission III. (*Brit. Weld. J.*, 1957, **4**, Dec., 556). A torsion test for welds in mild steel is specified.

**Weldability of Steels Having a Combined Ni-Cr-Mo Content of 3% and up to 0·28% C.** V. Pilous. (*Zváranie*, 1958, **7**, (1), 8–15). [In Czech]. Results of a detailed study of weldability and welding technology are given.

**On the Mechanism of Intercrystalline Crack Formation Near the Seams of Fusion Welding.** B. I. Medovar. (*Zhur. Tekn. Fiz.*, 1957, **27**, (7), 1571–1574). [In Russian]. In welding steel with a copper electrode, the penetration of copper into the base metal can be observed. The spreading of copper occurs only along the grain boundaries. It is suggested that the penetration of the low-melting point liquid from the weld reservoir is the result of adsorption migration and of the adsorption lowering of strength—the Rebinder effect.—L. H.

**Fabricating Continuous Galvanized Steel. II.** E. W. Horvick. (*Steel Proc.*, 1958, **44**, Feb., 92–94, 103–104). Uses of continuous galvanized strip, its soldering and welding, degreasing and finishing are reviewed.

**Experimental Investigation into the Mechanism of Joining by Soldering.** J. Colbus. (*Schweissen Schneiden*, 1958, **10**, Feb., 50–54). The mechanism of welding and joining during soldering and brazing is discussed. Tests were made with steel-to-steel, cast iron-to-cast iron and other metals, and temperature measurements were carried out. "Working temperature" and "wetting temperature" are defined.—U. E.

**Soldering Stainless Steel.** (*Products Fin.*, 1958, **22**, Mar., 59–60). A note on surface preparation and flux composition.

*J.I.S.I.*, 1958, **189**, Aug., 371*a*. The abstract **"Ferritic-Austenitic Weld Joints in Thermal Power Stations."** F. Zimmer, (*Eng. Dig.*, 1958, **19**, Mar., 99–102). was derived not from the *Rev. Met.*, but from the *Revue M de la Mechanique*. The original article surveys ten years' investigations on the subject and includes a large theoretical section and discussion. Effects of strain, diffusion and stress-corrosion are included and methods of minimizing these are given. Many types of joints are illustrated and special tests made on them are described.

## MACHINING AND MACHINABILITY

**Behaviour of Cutting Fluids in Reaming Steels.** L. V. Colwell and H. Branders. (*A.S.M.E.*, 1957, 57-A–168, pp. 5). Considerable rubbing is demonstrated on the margins of reamers, which is intensified by more effective lubricants. Torque requirements increase, hole size and surface roughness decrease with more effective lubricants.

**A Method for Studying the Behaviour of Cutting Fluids in Wear of Tool Materials.** L. A. Colwell. (*A.S.M.E.*, 1957, 57-A–160, pp. 5). Studies with a new friction and wear machine are described. Sharp discrimination between additives was achieved.

**Calculation of the Stages of Plastic Deformation of Metal in Cutting.** A. M. Rosenberg and G. L. Kufarev. (*Vestnik Mashin.*, 1958, (6), 49–52). The deformation in cutting is a simple displacement arising from compression. Three formulae can be derived from the data available of which only one gives exact results corresponding to practical conditions. The other formulae give results between 4% above to 20% below actual figures.—L. H.

**Facing Girders is Outsize Milling Job.** (*Metalw. Prod.*, 1958, **102**, Sept. 19, 1654–1655). A description of a new machine built at Halifax with a 6-ft vertical traverse and a feed of 30 and 60 or 40 and 80 in. /min. vertically and horizontally.

**Assumptions for Quantity Finishing from the Materials Side.** P. Wiest. (*V.D.I.-Z.*, 1958, **100**, June 1, 691–697). It is pointed out that for mechanized quantity finishing processes, uniformity of material and of the treatment of the semi-finished material is essential for good results. This principle is discussed under the heading of various treatments such as hot and cold working and machining.

**Can Ceramic Cutting Tools Replace Hard Metals** F. Kölbl. (*Planseeber*, 1958, **6**, Aug., 48–66). The manufacture of ceramic materials for cutting tools is described in outline, their properties are discussed and compared with those of other tool materials, particularly hard metals.

**Electro-Erosive Metal Machining—Theory and Practice of a Matured Process.** E. W. Ullmann. (*Maschinenwelt Elektrotechnik*, 1958, **13**, May, 188–191). Several examples of the use of electroerosive machining techniques are described and illustrated.

**Reducing Electrode Wear in Spark Machining.** W. Ullman. (*Metalw. Prod.*, 1958, **102**, Sept. 12, 1608–1614). The mechanism of spark erosion is discussed and types of generators and spark gap control are considered and the effects of the material, properties and shape of the electrode are reviewed.

## CLEANING AND PICKLING

**Metal Finishing in Czechoslovakia.** D. O. Bartl and O. Mudroch. (*Electropl. Met. Fin.*, 1958, **11**, Aug., 263–267). A general survey of the work carried out by the Czech Research Institute of Corrosion and Metal Finishing on electroplating, various aspects of metal cleaning and coating by spraying and other methods, metallic corrosion and the deterioration of plastics.

**Measurement of the Temperature Distribution in a Plate for the Determination of the Impingement Intensity in Shot-Blasting.** H. Püllenberg and U. Plöger. (*Stahl Eisen*, 1958, **78**, Aug. 21, 1203–1206). The correct adjustment of shot-blasting machines requires the measurement of the temperature distribution in the workpiece. The authors describe a novel method for obtaining the temperature distribution curve in plates using minature thermocouples. Using this device optimum descaling could be achieved by stepwise adjustment of the shot-blasting machine.

**Polishing Special Steels.** (*Aciers Fins Spec.*, 1958, July, 18–20). The article commences with a brief outline of the mechanism of metal polishing, and goes on to deal with aspects affecting special steels—these are generally harder and more resistant to abrasion; at the same time they are tougher, and have a high coefficient of friction and low thermal conductivity. The practice adopted in view of these factors is described.

**Contribution to the Study of the Anodic Dissolution of the Austenite of a Rimming Steel in the Course of Electrolytic Polishing in Fused Salts.** Z. Wojcik. (*Met. Corr. Ind.*, 1958, **33**, Apr., 163–170). The results are described of the microscopical examination of specimens of mild steel austenized at 1100° C and polished electrolytically at 900–950° C.

**The Dollars and Sense of Pickle-Liquor Treatment.** J. S. Joseph and E. T. Culver. (*Iron Steel Eng.*, 1958, **35**, March, 112–120). The paper discusses the problem of control and disposal of waste pickle liquor, particularly where large volumes accrue from continuous pickling of steel strip and where a multiplicity of lines may be employed using from 30–40 lb of acid per ton of steel pickled. (7 references).—M. D. J. B.

**Stream Pollution Reflects Operating Practices.**—J. E. Kinney. (*Iron Steel Eng.*, 1958, **35**, April, 97–100). The author defines pollution in the industrial sense of the word, namely waters carrying oils, solids, cyanides, metals and acids from industrial processes, and discusses the means of solving the stream pollution problem.—M. D. J. B.

## PROTECTIVE COATINGS

**"Fido" the Watch Dog.** B. T. Neill and K. E. Langford. (*Electropl. Met. Fin.*, 1958, **11**, Aug., 269–271). The construction and operation are described of an electronic safety device for application to periodic reverse plating baths left running overnight without supervision.

**A Study of Cadmium-Tin and Zinc-Tin Alloy Electrodeposits.** B. Cohen. (*Plating*, 1957, **44**, Sept., 963–968). Pretreatment and experimental plating procedures are described and various corrosion test results are given. The Cd-Sn coating gave particularly good results.

**A New Method of Obtaining Refractory Aluminium Coatings on Cast Iron by Metal Spraying.** T. Drazkiewicz and S. Gebalski. (*Prace Inst. Mech.*, 1955, **4**, (14), 42–67). [In Polish]. Positive laboratory results of obtaining refractory Al coatings on low-C steel by metal spraying have started the investigations on using this method on cast iron. The results of refractoriness tests at temp. 400°, 700° and 900° C are tabulated. The detailed researches show that metal spraying is suitable as a technological method of obtaining refractory Al coatings on cast iron, especially on spheroidal cast iron.

**Plastic Properties of Sprayed Aluminium Diffusion Coatings on Steel.** S. Gebalski. (*Prace Inst. Mech.*, 1957, **6**, (20), 3–31). [In Polish]. The author gives a detailed account of an experimental investigation of the properties of Al diffusion coatings sprayed on technical iron (0·005% C), low-carbon (0·15%–0·20% C) and medium-carbon steels (0·48% C).

The experiments were carried out at room temperature and at 800° C–1200° C, while the time of heating varied from 2 min to 5 h. The behaviour of diffusion coatings during bending, twisting, compression and in tension has been studied and the effect of the thickness of the sprayed layer on plasticity and refractoriness of the diffusion coatings has been observed.

**Phosphoric Acid Treatments for Steel. I. The Nature of Coatings Produced by the Action of Phosphoric Acid on Steel.** M. Donovan, J. W. Scott and L. L. Shreir. (*J. Appl. Chem.*, 1958, **8**, Feb., 87–96). Mild steel sheet was degreased, pickled in HCl, washed and dried. Without this treatment the surface was resistant to phosphoric acid. Coating was then carried out by soaking for various times and temperatures in $H_3PO_4$ and allowing to dry for 24 h. The weight and composition of the coating and the effects of oxygen were examined.

**The Selection and Use of Phosphate Coatings as Pre-treatments for Organic Finishes.** R. F. Drysdale. (*Electropl. Met. Fin.*, 1958, **11**, Feb., 51–55). A general review.

**Accelerators and Their Function in Phosphating Solutions.** L. O. Gilbert. (*Proc. Ann. Conv. Amer. Electroplaters' Soc.*, 1957, 73–78; discussion 162–163). Addition of heavy metals, electrolytic polarization and the use of organic depolarizers are reviewed and the mechanism of action is discussed. Effects of nitrate and nitrite are described.

**Choice of Testing Methods for Phosphate Coatings.** J. Weber. (*Prace Inst. Mech.*, 1956, **5**, (16), 19–24). [In Polish]. A short account of phosphate coatings and a survey of methods of accelerated and laboratory tests of phosphate coatings is given. The author describes in detail his own investigations and discusses the results. It is shown, from the drop test (according to Akimov), that the degree of corrosion (see Schuster and Krauser) and the size of the surface pores are interrelated.

**Metal Finishing in the Mining Industry.** (*Ind. Fin.*, 1958, **10**, Mar., 37–39). Bituminous and epoxide materials, wrapping tapes and anticorrosive paints are briefly discussed.

**Processing and Finishing Zinc-Coated Sheet.** (*Ind. Fin.*, 1958, **10**, Feb., 41). Organic coatings have been developed of the stoving-enamel type.

**Use of the "Colarit" Process for Spraying Coal-tar Enamel to Protect [a Hydro-Electric Installation] in Turkey.** H. Stern. (*Werks. Korr.*, 1957, **8**, Dec., 736–738). A practical account is given of the protection of the steelwork at a large hydro-electric station in Turkey. The procedure consisted in sand-blasting the steel, warming its surface by means of spraying pistols, applying a thin tack coat of coal-tar pitch solution, by hand or spray, and finally spraying on at 250° C a coat of coal-tar enamel 2 mm (80 mils) thick. The coating materials complied with specifications C203–51 and C204–51 of the American Waterworks Association.—J. C. H.

**Plastic-Coated Steel.** J. Summers and Sons Ltd. (*Auto. Eng.*, 1958, **48**, Jan., 17–18; *Ind. Fin.*, 1958, **10**, Feb. 37). A description of "Stelvetite" polyvinyl chloride coated sheet is given. The bending, forming, drawing and welding properties are shown.

**Oxidation of Iron Pretreated for Porcelain Enamelling.** L. E. Fussell and R. L. Hadley. (*J. Am. Ceram. Soc.*, 1958, **41**, Mar. 1, 81–88). Thickness of oxide layers was measured as a function of temperature to compare the effects of sulphuric acid pickle, nickel flash and phosphating. The Ni-flashed plate showed lower oxidation rates and the phosphated specimens significantly roughed interfaces between metal and oxide. X-Ray and optical methods were used.

**A Study of the Oxidation of Steel Plate as Related to Wettability and Adherence of Porcelain Enamel.** H. P. Still jun. (*Amer. Ceram. Soc. Bull.*, 1958, **37**, Jan., 22–26). An attempt to devise a rapid test to predict enamel behaviour is described. Adherence was correlated with oxidation but wettability was unrelated to both.

**Ceramic Finishing: Whirlpool Instals Continuously Recirculating System.** (*Products Fin.*, 1958, **22**, Mar., 50–54). A spray application plant for enamel coatings is described.

## CLAD SHEET AND HARD FACING

**Plate Clad with Special Steels.** H. Lejay. (*Rev. Gén. Méc.*, 1958, **42**, May, 273–278). The procedure adopted to line plate with stainless and other special steels is described, the process being suitable for medium and heavy plate. The alloy steel is tack welded between two plates, the material rolled together, and the clad plates afterwards separated. To promote adhesion, the plate is first plated with nickel where the alloy steel to be used contains chromium.

**The Development of the Production of Nickel and Aluminium Clad Steel.** M. Žídek. (*Výzkumné Práce z Oboru Železa a Oceli, State Publ. House of Tech. Lit., Prague*, 1958, 27–45). [In Czech]. Details are given of researches which led to the development of the production of steel strip made of low carbon rimming steel, clad on one side with Ni and on the other with Al. The maximum strip width was 16 cm, the thickness 0·015 cm. This type of strip was found satisfactory in applications in vacuum technology. The cost is about half of that of similar strip made of Ni.—P. F.

**Use and Application of Clad Steel, Metallic Liners and Overlay Deposits in the Manufacturing of Vessels for the Chemical Industry.** L. Capel and C. Nederveen. (*Lastechniek*, 1958, **24**, July, 136–149). A review of the use of clad steel, metallic liners and overlay deposits to combat corrosion in the chemical industry, with particular reference to welding problems encountered and their solution.

**The Lining of a Steel Pressure Vessel with Titanium.** F. F. Ates. (*Lastechniek*, 1958, **24**, July, 131–136). A detailed description is given of the welding of a Ti lining into a pressure vessel intended for the petrochemical industry. The work was done in panels, using the Argonare process.

**Surface Boronising of Low Alloy Steels.** V. D. Taran and L. P. Skugorova. (*Metallov. Obra*, Met., 1957, (6), 43–47). The quality of the boronized layer depends on the temperature of the process of electrolytic boronizing and the holding period. If the boronizing period is raised to over 6 h and bath temperature to above 950° C, the boronized layer begins to manifest brittleness. Optimum temperature for boronizing appears to be 950° C and the optimum period, six h in the case of the group of 0·8 to 1·1%Cr, 0·2 to 1·8%Si and 0 to 3%Ni steels investigated.—R. S.

## POWDER METALLURGY

**Powder Metallurgy Comes of Age.** R. Talmage. (*S.A.E.J.*, 1958, **66**, July, 44–47). The properties of sintered steel in comparison with cold-drawn 1020 steel are given, and the production and advantages of sintered steel gears are set out.

**Sintered Iron Base Alloys for Machine Parts. III, IV. Properties of Sintered Iron–Copper Alloys. I, II.** T. Watanabe, Y. Fujii and E. Hayashi. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, Oct., 544–548; Nov. 629–634). [In Japanese]. Dimensional changes can be controlled by control of particle size of the iron powders. The mechanical properties are much improved by modification of pore structure and by compact densification and alloying. Mechanical properties are improved by increase in Cu, but not so much with repressing as with single pressing. In single pressing, reduced Fe powder gives better products than electrolytic but the reverse is true for re-pressing.

**Complex Powder Metal Parts Call for Higher Density.** S. Bradbury. (*Iron Age*, 1958, **181**, June 5, 124–125). The advantages of multiple motion presses are described. Complex parts of uniformly high density can be pressed and carefully ejected.—D. L. C. P.

**Studies on Heat Treatment of Oil-Containing Sintered Ferrous Metals. I.** H. Nisimura. (*Suiyokwai-Shi*, 1956, **13**, Dec., 135–138). Physical properties and microstructures of carburized sintered iron showed differences between specimens; these are related to the sintering pressure, carburizing atmosphere and other factors.—K. E. J.

**Weakly Magnetic Materials on an Fe–Si–Al Base.** G. Rassmann, E. Lindner and R. Wittig. (*Neue Hütte*, 1958, **3**, June, 365–371). This article discusses the present state of development and application of the Fe–Si–Al alloys, temperature dependence of permeability, and the manufacture of sintered alloys of low coercive strength.

[**Powder Metallurgy**]. **Various Tooling Principles.** (*Höganas Iron Powder Handbook*, Chapter D.20, pp. 42). **Tool Design** (D.30, pp. 66). **Tool Making, I and II.** (D.35, pp. 122). **Sizing and Coining.** (D. 40, pp. 26). **Metal Powder Presses.** (D.50, pp. 36). These are sections on loose leaves for insertion into the handbook.

## PROPERTIES AND TESTS

**The Material Testing Machines of the 5th European Machine Tool Exhibition.** H. Mintrop. (*Werkstattstechn. Maschinenbau*, 1958, **48**, Jan., 41–50). A detailed account, illustrated with photographs, is given of the various types of materials testing equipment shown at the 5th European Exhibition.

**Relation between the Quality of Tube Round made from a Rimmed Steel Ingot and Inner Surface Defects of the Steel Tube. I. Tentative Method for the Estimation of Quality of Tube Round.** N. Eguchi. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 982–983).—K. E. J.

**Influence of Naturally Alloyed Iron Additions on the Mechanical Properties of Castings at Higher Temperature.** N. G. Girshovich and I. Ya. Ioffe. (*Lit. Proizv.*, 1955, (7), 1–4). [In Russian]. The specimens under investigation were Diesel engine parts. Tests for strength are described on specimens cast from different qualities of such irons, which have been subjected to instantaneous loads and to stresses of long duration between 150 and 10 000 hours at temperatures between 350–600°.

The findings have contributed to the preparation of diagrams and tables for the optimum quality and give the corresponding proportions of naturally alloyed iron which is to be added to the charge in each particular case.—L. H.

**Stress-Strain Relationship for Grey Iron.** E. M. Stein. (*Trans. A.F.S.*, 1956, **64**, 214–216). Typical curves for medium and low strength irons are given, and show that they have not a true modulus of elasticity. Ultimate strength plotted against strain, however, gave a definite relationship suggesting interpretation in terms of percentage ultimate strength.

**The Mechanical Properties of the Ferrite Crystal.** N. P. Allen. (*JISI*, 1959, **191**, Jan., 1–18). [This issue].

**The Mechanical Behaviour of Low Temperature Steels.** A. Galbiati. (*Acciaio Inossid.*, 1957, **24**, Sept.-Dec., 119–132). [In Italian]. Deformation under stress is examined in low temperature steels and the effects of lowering temperature on the mechanical characteristics of these steels are studied. Technical and statistical data on experiments are given.

**Electric Resistance-Wire Strain Gauges: The Influence of Differences Between the Sensitivity Factor k of Active and Dummy Gauges on the Exactness of Measuring.** Z. Rolinski. (*Prace Inst. Mech.*, 1956, **5**, (16), 58–60). [In Polish]. Analysis of errors in measuring, caused by the difference between the sensitivity factor k of active and dummy gauges is given. The calculations show that for established conditions, the difference in the case of an Al-base ought to be zero, on brass or steel the difference may amount to 0·03 and 0·05 respectively.

**Application of the Method of Resistance-Wire Strain Gauges to the Measurement of Stresses and Forces in the Processes of Plastic Treatment of Materials.** Z. Rolinski and O. Leszkiewicz. (*Prace Inst. Mech.*, 1957, **6**, (19), 13–29). [In Polish]. The author discusses the principles of dynamic stress measurement with the aid of resistance-wire strain gauges. He analyses the means of measuring and recording the forces appearing in the processes of deformation treatment of metal parts as a function of time. He describes in detail the apparatus used in his investigations and the measuring methods applied.

**The Magnitude of the Nondilatational-Strain-Indicator Reading for Large Deformations.** Z. Marciniak. (*Prace Inst. Mech.*, 1956, **6**, (18), 1–12). [In Polish]. The author discusses the equation of the indicator for small strains:

$$\epsilon_H = \frac{\sqrt{2}}{3}\sqrt{(\epsilon_x-\epsilon_y)^2+(\epsilon_y-\epsilon_z)^2+(\epsilon_z-\epsilon_x)^2+\tfrac{3}{2}(\gamma^2_{xy}+\gamma^2_{yz}+\gamma^2_{zx})}$$

The reading of the indicator for large deformations is based on an energy definition. A geometrical interpretation is given. The relation between the strain indicator and the components of the strain-velocity tensor are discussed, taking into account the rotation taking place during straining.

**The Effect of Alloying Elements on the Hot-tensile Properties of Iron Oxide Scale.** K. Kajikawa. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 990–991). [In Japanese]. The effect of the amount of alloying element (0–1·4%) on the tensile strength of oxides of alloys of iron with Ni, Si and Mn is investigated. The relationships between temperature and the tensile strength of oxides of iron with Ni, Si, Mn or Cr are shown. The tensile strength and elongation of electrodeposited iron are compared with those of its oxide at high temperature.—K. E. J.

**On the Flakes Appearing in Low-Carbon Killed Steel.** O. Otu. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 997–998). The flakes are those appearing in the fracture of test-pieces, and the relationships between tensile strength and elongation and reduction of area are shown for flaky and non-flaky specimens. The appearance of flakes is associated with low carbon (<0·20%) at melt-down, low ore additions and low refinement times (<13 min), and also with large bar sizes (19–25 mm).—K. E. J.

**Constant Head Rate Testing Machine with Instantaneous Speed Change.** M. J. Manjoine, E. T. Wessel and W. H. Pryle. (*Amer. Soc. Test. Mat. Bulletin*, 1957, Dec., 31–35). Changes in direction or speed over a range of $1 : 10^6$ can be made.

**A Modified Bend Test for Hardened Tool Steels.** G. Hoyle and E. Ineson. (*JISI*, 1959, **191**, Jan., 54–65). [This issue].

**The Mechanical Testing of Thin Mild-Steel Sheet and Strip.** M. L. Hughes. (*Sheet Metal Ind.*, 1958, **35**, Jan., 5–13). Production and testing are reviewed and descriptions are given of hardness, micro-indentation, bend tests, springiness variations, springback, cupping and tensile tests and the evaluation of surface finish. (17 refs.). Other topics were touched on in discussion. (14–18).

**Constructional Calculation of a Shrunk Fit in the Case of a Massive Steel Peg.** W. Korewa. (*Zeszyty Naukowe Politechniki Lodzkiej, Mechanika*, (4), 1956, **10**, 67–87). [In Polish]. Practical methods for graphical calculations of a shrink fit enabling the choice of the best sort of construction are given for the following three cases:

a—cast iron hub
b—steel hub—pressed fit
c—steel hub—shrunk fit

Ways of calculating the resulting stresses are shown.

**Determination of Optimum Hot-Working Temperatures.** V. Hašek. (*Strojírenství*, 1957, **7**, (12), 911–914). [In Czech]. Experimental equipment for the torsional plastic deformation of metal rods at elevated temperatures, as designed by the author, is described. It is shown that the temperature-dependence of the number of twists sustained up to fracture provides a satisfactory guide to the ductility in hot working operations. A plain carbon steel, two low-alloy steels and an 18/8 Cr–Ni steel were used in the experiments; the temperatures ranging from about 1000° to just under 1400° C.—F. F.

**On the Study of Impact Testing.** A. Vetiška and A. Till. (*Hutn. Listy*, 1958, **13**, (1), 37–41). [In Czech]. The deformation characteristics of several steels during impact testing were studied. The forces, transmitted by a piezoelectric crystal built into the hammer, were recorded on the screen of a cathode-ray oscillograph.—P. F.

**Impact and Fatigue Properties of Ductile Cast Iron.** C. F. Walton. (*Mach. Design*, 1958, **30**, Jan. 23, 128–131). Effects of structure, composition and heat treatment are shown.

**Some Considerations on the Notch Toughness of Steels.** H. Susukida. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 897–899). Notched slow bend test results and transition and energy-temperature curves are given for two groups of steel (specimens being taken from different parts of the plate), whose chemical composition and mechanical properties are also given.—K. E. J.

**Some Factors Affecting the Notch-Impact Properties of Carbon Steels. Influence of Aluminium and Nitrogen in Steels and Strain Ageing.** H. Gondo. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 894–897). Data are given for the effects of Al and $N_2$ on the transition points and strain-ageing of various converter and O.H. steels (analyses given).—K. E. J.

**On the Necessity of Fundamental and Widened Research in Strength of Materials.** M. V. Zaustin. (*Amer. Soc. Test. Mat. Bulletin*, 1957, Dec., 52–61). A general discussion of the strength of metals under various conditions of strain and the need for more data is presented.

**The Need for Specific Sheet-Metal Testing Methods for Deep Drawing and Forming.** G. de Witte. (*Sheet Metal Ind.*, 1958, **35**, Jan., 19–20). A lecture.

**A New Method for Testing the Deep Drawing Properties of Sheet Metals.** Z. Wisniewski. (*Prace Inst. Mech.*, 1954, **4**, (13), 53–63). [In Polish]. The new method is based on forming cups in four successive draws without interstage annealing. The application of this method is considered easy and the absence of failures confirms the deep drawing properties.

**Study on the Hot-workability of Steels. VI. High Temperature-High Speed Tension Tests of Austenitic Stainless Steel.** T. Morishima. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 986–988). The testing machine is described, and results from it are compared with those from hot torsion tests at temperatures of 1100°, 1200° and 1300° C on 18–10, 18–12–Mo, 18–12–Ti and 18–12–Cb steels.

**Effect of the Conditions of Hot-Cold Working on the Mechanical Properties of 19 : 9 D.L.** O. Ochiai. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1013–1015). The effects of working temp. (750–900° C) and amount of working (up to 15%) on hardness, impact, tensile, and creep-rupture properties are shown.—K. E. J.

**Effect of Cold Work on the Properties of High Alloy Steel.** M. A. Balter. (*Metallov. Obrab. Met.*, 1956, (5), 33–40). Although cold working was effective as regards the transformation of retained austenite, it did not ensure optimum impact strength, fatigue properties and wear resistance of case-hardening Cr-Ni-V steel. Full tempering at 650° C after cementation (prior to quenching), which ensures a homogeneous structure of fine acicular martensite with uniformly distributed dispersed particles of residual austenite, reduces tensile stresses in the case. The nature of the distribution of the residual austenite and the structure of the martensite affect the fatigue strength, wear resistance, ductility and impact strength of this steel. The amount of residual austenite, of itself, is not a complete guide to physical and mechanical properties.—R. S.

**Limiting Stresses in Steel under Static Loading and in the Case of a Steel which is Ductile up to Failure.** L. Baes. (*Acier-Stahl-Steel*, 1958, **23**, March, 125–134). [In French]. The paper makes an advanced study of three-dimensional stresses in solids with ductile behaviour up to the point of failure. The new Belgian specifications for rolled, forged and cast steels are examined in the light of theoretical and practical considerations governing the behaviour of steels. A method of graphical representation of stress conditions is described.—M. D. J. B.

**Effect of Chemical Composition and Microstructure on the Workability of Chromium Irons.** M. M. Levitan. (*Metallov. Obrab. Met.*, 1956, (2), 50–52). The workability of high carbon steels (2 to 3%C), and Cr alloys (0 to 21%Cr) is reviewed. It is shown that when the carbon content is increased, workability deteriorates. Cr improves workability. Optimum workability was found in an alloy containing 13 to 21%Cr, 1·9 to 2·1%C and 0·3 to 0·7%Si. The workability of high carbon alloys containing Cr improves with an increase in the amount of primary carbides of the trigonal type.—R. S.

**Metal Fatigue.** N. Thompson and N. J. Wadsworth. (*Advan. Phys.*, 1958, **7**, Jan., 72–169). A comprehensive review of experimental methods, metallography, mechanical properties, fatigue limit, notch effect and crack propagation, X-ray studies and surface effects and experiments on annealing and effects on diffusion, with discussion is given.

**Fatigue of Metals—II.** C. E. Phillips. (*Tek. Uke.*, 1958, **105**, Feb. 20, 169–175). The author discusses influence of notches and stress gradients and compares theories of fatigue failures with recent findings. Design considerations with special reference to screwed and pin joints are described in the second half of this part of the article.

**Fatigue Testing Machine for Hot Sheet.** K. W. Mitchell and H. King. (*Engineering*, 1958, **185**, Mar. 28, 402–404). A machine developed at the Fulmer Research Institute to give a constant deflexion up to $\pm 0{\cdot}4$ in. with a load limited to $\pm 4$ lb is described and a typical test diagram for a ferritic steel at 650° C is given.

**Fatigue Fracture Criterion in Metals Based on Dislocation Theory.** T. Yokobori. (*Technology Reports of Tôhoku Univ.*, 1957, **22**, (1), 51–61). The criterion is developed mathematically and applied to alternating stress conditions. The phenomenon of cracking is interpreted. (21 refs.).

**Fatigue Tests on Basic-Bessemer and Open-Hearth Steels.** A. Funck. (*Acier Stahl Steel*, 1958, **23**, Jan., 19–24). A range of steels representative of those in current use for steel structures were compared. No systematic difference was found between basic Bessemer and O.H. steels. The difference between elastic and fatigue limits increases with the former according to a formula given. Fatigue tests of $10^6$ cycles were found to be sufficient to distinguish between different steels for tendency to failure.

**Fatigue Strength of E-I 723 Steel at High Temperatures.** B. I. Aleksandrov. (*Metallov. Obrab. Met.*, 1956, (12), 27–33). E-I 723 steel contains C 0·20, Cr 2·3, Mo 0·93, V 0·45, Mn 0·74, Si 0·28, S and P 0·010% each. In the temperature range of 20° to 500° C, E-I 723 steel possesses high fatigue strength after normalizing and full temper. Above 500° C, however, there is an abrupt drop in fatigue strength. At temperatures up to 400° C, this steel fails because of fatigue. At 500°, 535° and 575° C, with cyclic stressing in air, failure of the steel is due to corrosion fatigue. Surface work hardening by rollers improves fatigue strength.—R. S.

**Effect of Various Types of Treatment on the Fatigue Strength of EI 434 Steel.** B. I. Aleksandrov and A. P. Shishkova. (*Metallov. Obrab. Met.*, 1956, (8), 20–27). At 20° and 650° C, EI 434 steel has a high fatigue strength (31·5 to 25 kg per sq mm) after quenching and stabilizing. When the temperature is raised to 750° C, the fatigue strength is markedly reduced (18 kg/mm²). Stabilization conditions exert no marked effect on fatigue strength at 700° C. The fatigue strength of notched specimens (when the notch is cut) is 80% higher at 20° C and 22% higher at 650° C compared with that of specimens in which the notch has been produced by a grinding wheel. Surface work-hardening by rollers also increases the fatigue strength of parts.—R. S.

**The Effect of Surface-Roughening Applied Prior to Metallising on the Fatigue Strength of Components.** V. Linhart. (*Strojírenství*, 1957, **7**, (12), 917–921). [In Czech]. The effect of various types of surface roughening on the fatigue strength of a 0·4%C, 1·96% Cr, 0·17%V steel before and after metallizing was investigated. If the metal surface was not grooved for the retention of the metallizing layer then shot and sand-blasting did not significantly affect the fatigue strength of the metal either before or after metallizing. Otherwise, or on using a metal arc for roughening, the fatigue strength declined by 25–50%.—P. F.

**Analysis of Fatigue Failures.** M. E. Niezgodzinski. (*Zeszyty Naukowe Politechniki Lodzkiej, Mechanika*, (4), 1956, **10**, 113–124). [In Polish]. Conditions in which fatigue foci arise are discussed. The influence of the value and distribution of stresses on the shape of the fatigue fissures and on the system of front curves are described. Photographs of most types of characteristic fatigue failures of straight and crank shafts are given and analysed in order to determine the direct causes of the fatigue foci.

**Influence of Arsenic on the Fatigue Strength of Rails.** L. M. Shkolnik. (*Metallov. Obra. Met.*, 1956, (2), 14–21). The harmful effect of As becomes apparent from a content of 0·15% upward in tests on rails, and is intensified as the endurance limit is reached. The effect of As on the fatigue strength of specimens cut from rails is much less and it is therefore concluded that the deleterious effect of As is at the surface. The effect of C and Mn on tensile strength depends on the level of the acting stresses and increases with reduction in stress and with increase in the period of cyclic loading. The investigations showed that the replacement of long time fatigue tests by short time tests can often lead to incorrect conclusions.—R. S.

**Fatigue Resistance of Crankshafts of Iron with Nodular Graphite and of Steels, with Surface Tempered Journals.** I. V. Kudryavtsev and N. A. Balabanov. (*Lit. Proizv.*, 1955, (11), 15–17). [In Russian]. Under correct conditions of production the surface hardening, or cold hardening of straight shafts ensure freedom from trouble. In crank shafts, considerable distortion may arise either due to the surface hardening of the journals or to the cold hardening of the chamfers.

For the elimination of these difficulties one can use with advantage the method of local cold hardening of the crank shafts in the zone of the journals, method which has been in practical use for a long time.—L. H.

**The Fatigue Strength of Rolling Stock Components.** J. Němec. (*Strojírenství*, 1957, **7**, (12), 883–889). [In Czech]. Rolling stock components are frequently exposed to complex alternating stresses, including impact loading arising from discontinuities in the rails. The paper surveys the theoretical and practical bases for estimating the fatigue strength of components subject to such complex alternating stresses. Experimental work on periodic impact loading of copper and steel wires, done by the author, is reported, and the fatigue behaviour observed is analysed.

**Contribution to the Static Theory of Creep and Relaxation.** G. F. Lepin. (*Izvest. Akad. Nauk, Otdel. Tekhn.*, 1957, (9), 134–136). [In Russian]. In this work, carried out under the direction of I. A. Oding, the author attempts to deduce a general equation of deformation, and equations of creep and relaxation as particular cases. He bases these on the dislocation theory and static relationships.—S. K.

**Studies on Creep Rupture of Steels by Hot-Stage Microscopical Examination.** M. Tanine. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1007–1009). Creep-rupture results are given, and the failures are shown by the microscope to be caused by transcrystalline fracture transverse to twins, intercrystalline fracture, and fracture at the twin boundary.—K. E. J.

**The Kinetics of the Change in the Microstructure of Metals and Alloys during Creep during High-Temperature Stressing *in vacuo*.** M. G. Lozinskii and E. I. Antipova. (*Metallov. Obra. Met.*, 1955, (5), 9–14). Investigation of creep in austenitic carbon steel at 900° has made it possible to establish that the deformation process originates for preference at the grain boundaries of the austenite. At experimental temperatures, these boundaries are much less resistant than the centres of the grains.—L. H.

**Study of Heat-resisting Steel. XII.** E. Asano. (*Tetsu-to-Hagane*, 1957, **43**, 1132–1136). [In Japanese]. Some data on creep and age-hardening characteristics are given for Timken 16–25–6 samples containing various amounts of Ti or B. The addition of Ti improves creep strength, and addition of B in place of $N_2$ may be desirable to keep creep strength at the standard level.—K. E. J.

**An Investigation of Cracked A-212 Catapult Steam Receivers.** A. J. Babecki and P. P. Puzak. (*Naval Research Laboratory Report* 5058, 1957, pp. 12). Defects in aircraft carrier equipment are examined and corrosion and fatigue cracking structures are shown.

**Modified Jominy Test.** M. Kawai. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 975–978).—K. E. J.

**On the Hardenability of Self-Hardening Steels.** K. Kiyonaga. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 978–980). [In Japanese]. Various data are given for eighteen steels (analyses given).—K. E. J.

**The Measurement of Meyer Hardness and Young's Modulus on the Rockwell Test Machine.** D. K. Mackenzie and T. P. Newcomb. (*Brit. J. Appl. Chem.*, 1957, **8**, Oct., 398–399). Comparisons with Vickers hardness and published values of Young's modulus were made.

**Hardenability and Distribution of Hardness Throughout the Section of Quenched Steel Bodies.** M. E. Blanter and A. A. Gol'denberg. (*Metallov. Obrab. Met.*, 1956, (8), 12–20). After discussing the fundamental qualitative laws governing hardenability, the authors use a series of nomograms for calculating the distribution of hardness in relation to hardenability. The nomograms also show the relationships between the various characteristics of hardenability.—R. S.

**Hardness of Steel Quenched After Induction Heating.** I. Ya. Gorbul'sky. (*Metallov. Obrab. Met.*, 1956, (3), 34–37). With rapid heating, not only is the optimum quenching temperature raised but the temperature range in which the best quenched structure is obtained is extended. With induction heating, heat is generated directly in the surface layer of the metal and thus the speed of austenite transformation is substantially raised. The higher temperature also accelerates austenitizing. It is thought that the simultaneous effect of these two factors excludes any influence of initial structure on the hardness of steel. The authors found that when steel was induction heated and normalized, greater hardness was obtained than in quenched annealed specimens. A formula is given for determining the hardness of quenched steel in relation to carbon content.—R. S.

**Studies on Fire Cracks. I.** J. Watanabe. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 996–997).

**Effects of Plating High Tensile Strength Steels.** W. Beck and E. J. Jankowsky. (*Proc. Ann. Conv. Amer. Electroplaters' Soc.*, 1957, 47–52; discussion, 161–162). The adverse effect of $H_2$ introduced into C-ring specimens during plating is described. A simplified testing machine and test piece was developed and plating with Cd, Cr and Sn was examined, with various protective treatments before and after plating.

**On the Critical Ductile to Semi-Brittle Transformation Temperature in Structural Steel.** I. E. Kontorovich and B. M. Voshedchenko. (*Metallov. Obrab. Met.*, 1956, (5), 19–24). Diagrams showing the relationship between critical temperatures and the form of fracture in structural steels were confirmed by test data. Three regions of tempering temperature were determined, in which the steel is in a ductile, semi-brittle or brittle state.—R. S.

**Study on the Hot Workability of Steel for Special Shapes. I. Effect of Blank Materials. II. Effect of Copper and Tin.** A. Suseki. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 920–921; 921–923). Tests to investigate hot-shortness are reported for several types of rimmed steel and killed steel (produced by O.H. and arc furnaces). The influences of Cu and Sn on hot-shortness are examined.—K. E. J.

**Causes of Crack Formation in Cemented [Machine] Parts and Measures to Combat Them.** E. I. Malinkina. (*Metallov. Obra. Met.*, 1955, (4), 39–47; (5), 24–28). Cracks can form in the core below the cemented layer or superficially in the layer itself. The internal cracks appear for preference during tempering and can be eliminated by using fine-grained alloys which makes overheating unnecessary but which gives a hardness below 45$R_C$.

The superficial cracks appear on tempering at low temperature leading to the formation of a martensite structure throughout the whole cemented layer. More rapid tempering is of advantage in such cases.—L. H.

**Intercrystalline Fracture of High Speed Steel.** L. B. Getsov. (*Metallov. Obrab. Met.*, 1956, (8), 42–43). This defect, having the appearance of naphthalene, is encountered when tools are subjected to repeated quenching without intermediate annealing. It can be corrected by heating to 1140–1160° C and holding for three to eight minutes followed by cooling by holding for 15 to 30 min at 720–800° C and subsequent water quenching. This treatment is repeated and followed by annealing and finally the normal heat treatment.—R. S.

**Causes of Cracking in Superheater Tubes Made of EI 257 Steel.** L. P. Trusov, L. P. Nikitina and G. A. Tulyakov. (*Metallov. Obrab. Met.*, 1956, (5), 27–33). Boiler operational conditions and the nature of the occurrence of transverse cracks are stated. Microstructure examination, hardness measurements, determination of micro-mechanical properties, intergranular corrosion tests and fatigue tests at high temperatures were all used to ascertain the causes of cracking of some tubes and the absence of cracking in other tubes operating under similar conditions. It was found that the cracks had a fatigue origin and the main causes were lack of properly controlled operational conditions of the boiler during the running-in period which gives rise to repeated thermal stressing; condensation of chemically cleaned boiler feed water in the lower tubes; work-hardening of tubes caused by the cold bending of 55 mm diameter tubes during manufacture; incorrect heat treatment of the tubes.—R. S.

**Nature of Temper Brittleness.** N. N. Davidenkov. (*Metallov. Obrab. Met.*, 1956, (12), 36–40). Recent work on the mechanism

of temper brittleness is discussed in relation to the effect of the rate of impact stressing and bending on temper brittleness.—R. S.

**Mechanism of the Nature of Temper Brittleness.** N. V. Tolstoguzov. (*Metallov. Obrab. Met.*, 1956, (8), 28–30). Previous work has shown that the cause of a marked drop in impact strength when temper brittleness develops is a reduction in tensile strength. This was shown by tests in liquid nitrogen and liquid air and could not be observed at room temperature. The present work is an attempt to show the difference in tensile strength of smooth specimens of brittle and ductile steel at room temperature.—R. S.

**The Nature of Temper Brittleness in Pearlitic Steels.** V. I. Prosvirin and E. I. Kvashnina. (*Metallov. Obrab. Met.*, 1956, (2), 34–49). Temper brittleness of the secondary type which occurs in structural steels during slow cooling after tempering in the range 600 to 400° C or during prolonged heating in the range 500 to 550° C, in which temper brittleness develops, is connected with the formation of carbides of the cementite type in the boundary zones. The effect of Mo is discussed.—R. S.

**Temper-Brittleness of Low-Manganese Steel. I.** T. Iwata. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1061–1063).—K. E. J.

**Location of the Threshold of Cold Rupture in Temper Brittleness.** V. I. Prosvirin and E. I. Kvashnina. (*Metallov. Obra. Met.*, 1955, (3), 17–20). The appearance of temper brittleness in steels seems to displace the threshold of cold rupture to higher temperatures, but an increase in the duration of tempering at high temperatures displaces this threshold to lower temperatures. An increase, in the size of the grains again displaces it to higher temperatures.

A displacement of the cold rupture threshold is not observed in steels not subject to temper brittleness.—L. H.

**Thermal and Temper Brittleness of Ferritic (or Pearlitic) Steel.** P. B. Mikhailov-Mikheev. (*Metallov. Obrab. Met.*, 1956, (2), 23–33). An analysis of the temperature-impact curves for the brittle and ductile states of ferritic steels shows that this type of brittleness is a feature of cold shortness observed in all ferritic steels subjected to the notch impact test at low temperature. Under the influence of embrittlement processes, the impact temperature curves are displaced into the region of higher temperatures and brittleness begins to appear at room temperature. The causes of temper brittleness are associated with changes in the structure of the initial grain boundaries which occur during the processes of thermal brittleness and reduce the effect of mechanical work-hardening and chemical stability in the boundary regions. The physical nature of the processes causing such thermal brittleness is still not known in spite of such research. It is doubtful, however, whether the causes of thermal brittleness are to be found in the finely dispersed precipitated particles of some new carbide phase. A more likely suggestion is that thermal brittleness is due to segregation of atoms of dissolved elements at the grain boundaries without the precipitation of a crystalline phase from a solid solution and to the distortion of the crystal lattice of the solvent at the grain interfaces.—R. S.

**The Problem of Brittle Fracture in Structural Steel.** W. Felix and T. Geiger. (*Schweisstechn.*, 1957, **11**, Dec., 158–162). An investigation is described on the behaviour of steel during fracture due to notch impact testing. X-Ray diagrams of Schnadt samples are given which show the variation in the orientation of the crystals. It is concluded that the crystal orientation is decisive for the behaviour of the steel during deformation and fracture.—U. E.

**The Classification of Steel Structures According to Their Tendency to Brittle Fracture** K. Rühl. (**Schweissen Schneiden*, 1958, **10**, Feb., 39–44). The new German standards for steel are discussed in detail. The standards classify steels of the same strength and elongation according to notch impact strengths which vary.—U. E.

**On the Nature of Reversible and Non-Reversible Temper Brittleness of Structural Steel.** V. S. Mes'kin. (*Metallov. Obrab. Met.*, 1956, (7), 24–35). The three hypotheses forming the starting points of a theory of reversible temper brittleness are set out. These are: (1) that the processes causing temper brittleness occur in the layer of initial austenite grain; (2) that they occur in the alpha solid solution and not in the carbide phase; (3) that the boundary layer of the grain having a thickness of many atoms is enriched with the elements to a greater or lesser degree depending on conditions of prior heating, rate of cooling and the nature and speed of diffusion of the relevant elements. The effect of elements and, in particular, the complex effect of carbon on temper brittleness is discussed.—R. S.

**Effect of Heat Treatment on Blue Brittleness of Structural Steel.** G. I. Pogodin-Alekseev and M. M. Fetisova. (*Metallov. Obrab. Met.*, 1956, (8), 31–37). The location and extent of the temperature range in which blue brittleness develops do not depend on the type of heat treatment associated with recrystallization. It was found that the best heat treatment for reducing blue brittleness in 0·55% C steel was subcritical annealing at 550° C. The effect of different periods of annealing on properties associated with brittleness are discussed.—R. S.

**Influence of Some Processes of Heat Treatment on the Tendency to Delayed Failure of Steel with a Tensile Strength of 120 to 140 kg/mm².** L. M. Pevzner, V. E. Sadovskii, T. K. Zilova, S. S. Volkov and Ya. M. Potak. (*Metallov. Obrab. Met.*, 1956, (3), 5–14). The nature of delayed failure at room temperature under constant stress is described. One of the main factors is the state of the surface layer which is affected by mechanical and heat treatment. A bath composition is suggested to eliminate the danger of $N_2$ pick-up during bright annealing in the range 500 to 550° C. Data are presented to show the effect of various heat treatments on the tendency of Cr steel to brittle fracture when subjected to stress and $H_2$ attack.—R. S.

**Temper Brittleness of CrMnNi Steel in Relation to Further Alloy Additions.** M. P. Braun and E. E. Maistrenko. (*Metallov. Obrab. Met.*, 1956, (3), 28–33). The addition of 0·28%Mo to a steel containing 1·25%Cr 1·25%Mn and 1·5-2·5%Ni is much more effective in reducing the tendency to temper brittleness than the addition of 0·52%W. However, the addition of 0·70%W almost eliminates temper brittleness in these steels. The best combination of alloy additions to these steels is W + Ti or W + V + Ti.—R. S.

**A Qualitative Separation of Wear Factors.** L. Rozeanu and O. Preotescu. (*Eng. Dig.*, 1958, **19**, Mar., 102–105; from *Revue de Mécanique Appliquée*, 1957, **2**, (1), 115–127). Adhesive, Abrasive, Corrosive and Fatigue processes are distinguished.

**Wear of Metals.** F. T. Barwell. (*Wear*, 1957–58, **1**, Feb., 317–332). The mechanism of wear is discussed with particular reference to the intimate structure of the layers in contact. Studies on steel and the course of fretting corrosion are described. Fragmentation of pearlite is shown to be a factor in wear.

**Load Capacity of Globoidal Worm Gear Transmission with Cast Iron Crowns.** G. A. Revkov. (*Vestnik Mashin.*, 1957, (10), 35–38). [In Russian]. Experiments were carried out on crowns made of antifriction low-alloy sulphur iron, made in the cupola, which is easily manufactured under ordinary conditions. Its composition was: C 3·61; Si 2·5; Mn 0·79; S 0·10; P 0·12; Cr 0·14 Ni 0·28%; and free carbon (as graphite) 2·73%.

From considerations of the character of the change in load capacity as a function of speed and also of the change in the coefficient of friction, it follows that at slipping speeds greater than 1·2 m/sec the globoidal worm gear with a crown of antifriction iron CCHTs2 is less satisfactory in use in comparison with worm gear transmissions of nonferrous metal.—L. H.

**Calculation of the Resistance of Metals to Abrasion.** M. M. Tenenbaum. (*Zhur. Tekn. Fiz.*, 1957, **27**, (5), 1106–1117). [In Russian]. Experiments were made with a trihedral diamond pyramid with an angle of 65° between the equal facets and the central axis. The pyramid can be moved laterally at an angle of 15° with the surface of the polished section.

Results showed that the resistance to abrasion is chiefly determined by the geometry of the scratch rather than by its magnitude. The magnitude of the scratch diminishes slowly with the increase in hardness.

The formulation of an index of resistance has been evolved which can be used for the determination of the hardness of pure metals, of cementite and of layers of diffusion chromiding. It bears a linear relation to microhardness.—L. H.

**On the Abrasion Resistance of High-Chromium Cast Iron. II. Blades for Shot Table Blast.** J. Hara. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 904–905). The hardness of six specimens (compositions given) of varying Cr content after heat treatment at 1050° C and air cooling are given. The weight loss is shown to decrease as Cr content increases from 5% to 35%.—K. E. J.

**Investigating the Wear of Tools with Radioactive Isotopes.** N. F. Kazakov. (*Vestnik Mashin.*, 1954, (4), 50–56). This method makes it possible to speed up the investigation. It has important significance in determining the relative magnitude of the wear-resistance of tools under different conditions of cutting and of pretreatment of materials in work. It can also be used in investigations with different lubricating and cooling fluids.—L. H.

**Investigating the Wear of Cutting Tools by Use of Labelled Atoms.** E. P. Nadeinskaya. (*Vestnik Mashin.*, 1954, (4), 40–50). This method is superior to the micrometric method for investigating the relation between wear and speed of cutting because it excludes the influence of the time factor and the non-uniformity of tools and materials in work. It can also determine the wear linearly or by volume. It does not exclude the micrometric method.—L. H.

**Wear Resistance of Steel Hardened by Normal and Induction Heating.** G. A. Preis. (*Metallov. Obrab. Met.*, 1956, (3), 47–50). High-frequency hardening of 0·45% C steel doubles the wear resistance under conditions of dry or semi-dry friction at low sliding speeds but improves it by only 25% at high sliding speeds. High-frequency heating has no advantages over normal furnace heating when friction is encountered at high temperature. With rolling friction, high-frequency heating substantially improves wear resistance.—R. S.

**Wear Resistance of Steel Surfaces Impregnated with Carbide-Forming Elements.** G. N. Dubinin. (*Metallov. Obra. Met.*, 1957, Sept., 21–25). Carbon steels with 0·03–1·18% C with surface treatment with Cr, V, Nb, W, Mo and Mn were examined for wear resistance. Microhardness and the results of wear tests are given. Reasons for anomalies in the relationship between hardness and resistance to wear are advanced.

**New Methods of Increasing the Wear Resistance of Tracks with Open Bushings.** A. A. Maurakh. (*Vestnik Mashin.*, 1956, **36**, (12), 16–18). Tractor tracks are discussed. Boronizing of austenitic Mn steel castings and performance of heat-treated alloy steels on 1% C tool steel as bushings are reported and compared with boronized 0·45% C steel. An electrolytic boronizing process is described.

**Transfer of Tungsten Carbide to Soft Metals During Single-Traverse and Reciprocating Sliding.** J. Golden and G. W. Rowe. (*Brit. J. Appl. Phys.*, 1958, **9**, Mar., 120–121). Tracer studies of sliding on Cu and on mild and stainless steels are described.

**On the Conditions of Atoms in the Ferromagnetic Alloys of Iron, Cobalt and Nickel.** I. Ya. Dekhtyar. (*Izvest. Akad. Nauk. Seriya Fiz.*, 1957, (10), 1452–1458). The study

of the magnetic properties of alloys of the transition elements is one of the methods of finding their electronic structures.

The non-regular dispersion of ions in ferromagnetic alloys is found to be proportional to the square of the difference of the magnetic moments of the two components of the alloy.—L. B.

**Effect of Heat Treatment on the Magnetic Properties of Generator Sheets of About 2·5% Silicon Content.** M. Markuszewicz and G. Groyecki. (*Prace Inst. Ministerstwa Hutnictwa*, 1957, **9**, (3), 125–135). [In Polish]. The authors discuss different problems connected with the manufacturing of electrotechnical sheets of silicon steel such as: the phenomena of diffusion during heat treatment, the character and duration of the process, problems of low-reduction and cold-rolling of generator sheets. They investigated the influence of Si (up to 40%) on the diffusion of carbon in steel and the application of finishing reduction (up to 10%) followed by annealing on losses in sheets. The results are tabulated and illustrated with curves and photographs. It was established that the lowest losses are obtained after annealing at 800–960° C.

**Magnetic After-Effect in Silicon Steel Subjected to Heat-Treatment.** A. Smolinski and M. Zbikowski. (*Arch. Electrotech.*, 1957, **6**, (2), 300–303). [In Polish].

**Flux Waveform Control in Iron Testing.** J. McFarlane and M. J. Harris. (*Elect. Rev.*, 1958, **162**, Mar. 21, 521–523). A review of methods and apparatus with discussion.

**Indicating FeO-meter for Sinter.** T. Yatsuzuka. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 872–873). Results show a close correlation between the readings of an inductance-type meter and the FeO content of sinter.—K. E. J.

**Spin-disorder Effects in the Electrical Resistivities of Metals and Alloys.** B. R. Coles. (*Advan. Phys.*, 1958, **7**, Jan., 40–71). Iron is included with rare-earth and transition metals and alloys.

**Improved Methods for Testing and Grading Electrical Steel.** J. McFarlane, P. Milne and J. K. Darby. (*Elect. Rev.*, 1958, **162**, Mar. 21, 519–520). Loss measuring instruments are reviewed and the installation at Bilston developed by the G.K.N. Group Laboratory is described.

**The Influence of Production Technology on Power Losses in Transformer Steels.** J. Skála, M. Knotek and R. Vokáč. (*Hutn. Listy*, 1958, **13**, (1), 14–21). [In Czech]. An account is given of operational research on the factors in the production of transformer steels affecting watt-losses to an appreciable extent. The procedure and the statistical analysis of the results are discussed. The most important factor was found to be the silicon content in the melt. By means of the results obtained modifications in the production technology could be introduced, leading to a reduction in the average loss-factor of the sheet.—P. F.

**Crack Depth Measurement.** (*Can. Mines, Research and Special Projects Report*, 1957, July-Dec., 16). An improved indicator has been constructed using electrical conductivity for D.C.

**Some New Equipment for the Non-Destructive and Continuous Testing of Metals.** L. Beaujard and V. Husarek. (*Métaux-Corrosion-Indust.*, 1957, **32**, Dec., 459–468). Modifications of the electronic circuit of ultrasonic test equipment which increase the efficacy of the technique are described. The use of ultrasonic methods for the continuous routine testing of rounds and bars is reported. An automatic ultrasonic test using the PPI radar-type display which gives a panoramic picture of defects in a sheet is described. Magnetic methods of testing are also considered.—R. G. B.

**Ultrasonic Investigation of the Structure of Steels.** L. G. Merkulov. (*Zhur. Tekn. Fiz.*, 1957, **27**, (6), 1386–1391). [In Russian]. The experiments were made on structural carbon steels (0·15 and 0·4% C) and tool steel (1·2% C). These experiments show that this method is suitable not only for purposes of defectoscopy, but also for investigating structure, for example, the non-tempered structure of steels, the quality of forged articles, the measurement of the average grain size, etc. The method, based on the damping of ultrasonic waves in the metal, opens up considerable possibilities in this direction of research.—L. B.

**On the Ultrasonic Testing of Steel Slabs. I.** Y. Sasaki. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1073–1074).—K. E. J.

**Relation between Penetrating Characteristics of Ultrasonic Waves and Mechanical Properties on Rolling Stock Axles.** R. Inoue. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1074–1076).

**X-Ray Methods of Stress Determination in Metals.** Z. Ziolowski. (*Prace Inst. Ministerstwa Hutnictwa*, 1957, **9**, (5), 203–225). [In Polish]. A survey of X-ray methods of stress determination in metals and their theoretical basis is given.

A modified and rapid method of "oblique screened photographs" has been worked out by the author and the technique is described. Methods of measuring and calculation of stresses from diagrams by conventional and rapid methods are given. Results of stresses studied in railway lines are shown.

**Radioactive Isotopes in Engineering.** M. B. Neiman. (*Vestnik Mashin.*, 1954, (4), 34–39). Radioactive isotopes are now used extensively for determining discrepancies in dimensions, for the control of production and quality, for detecting internal corrosion, for following currents through pipes, for controlling fluids in tanks and calculating the rate of gas flow. They are used for fire alarms, for ventilation control and for the solution of a great many problems connected with improving quality, reducing waste and costs of production.—L. B.

**Inspection of Castings with [γ-Rays from] Radioisotopes.** B. B. Gulyaev and L. G. Demina. (*Lit. Proizv.*, 1956, Sept., 18–20). Suitable sources and the defects they reveal are surveyed, the methods used and their sensitivity are discussed and examples are shown.

**Effects of Neutron Irradiation on Metals and Alloys.** A. H. Cottrell. (*J. Brit. Nuclear Energy Conf.*, 1958, **3**, Jan., 50–67). A review of neutron and ion collision processes and their effects on metals. Mechanical properties of irradiated metals are noted including the embrittlement of steel. (95 refs.).

**Collection of Thermodynamic Data for Use by Iron and Steel Metallurgists.** M. Ray. (*Publ. IRSID*, 1955, B, (26), Nov., pp. 117). Tables are presented showing thermodynamic constants for metals and other elements, gases and salts of interest for the iron and steel industry. Curves of enthalpy and entropy as a function of temperature are added.

**On Heat Content and Heat Capacity of Pure Iron and Cast Iron.** I. P. Egorenkov. (*Lit. Proizv.*, 1955, (7), 20–24). [In Russian]. The unreliability of heat calculations in the foundry industry is due most of all to the uncertainties in the basic thermodynamic values; hence this study which aims at tightening up such calculations.

Expressions are derived in calories per gr. atom, among others, for the transition phase and are compared for cast iron and for pure iron.—L. B.

**Effects of Structures on the Changes of Internal Friction of Some Steels Caused by Absorbed Hydrogen.** M. Mizuta. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1057–1059).—K. E. J.

**The Effect of Nitriding on the Internal Friction of Steels.** J. Vodsedálek and R. Štefec. (*Hutn. Listy*, 1958, **13**, (1), 9–13). [In Czech]. The damping capacity of several austenitic and ferritic alloy steels subject to static stresses was studied in the nitrided and untreated conditions, and the effects of structural changes on the damping capacity are discussed. Nitriding leads, in general, to a slight decrease in damping capacity; the latter may also undergo changes while the steel is in use, particularly in the case of austenitic steels.—P. F.

**Steels at Elevated Temperatures with Special Reference to the Use of Molybdenum.** J. D. Murray. (*Alloy Met. Rev.*, 1958, **9**, Mar., 2–8). Corrosion resistance, fatigue properties, weldability, creep and ductility are reviewed.

**On the High-Temperature Strength of Low-Alloy Steels. IV.** S. Yamamoto. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1004–1006). Creep rupture results at 500° C are given for Cr–Mo–V and Ni–Mo–V steels (analyses given) after various specified heat-treatments.

**Time-Stability of Steel at High Temperatures.** H. Holdt. (*Jernindustri*, 1958, **39**, Feb., 29–33, 39–40). Modern methods lead to higher working temperatures, and steelworks are engaged on developing new steels which tolerate these. The author gives an account of creep strength—short and long term—and the results of experience in tests of different steels which are tabulated.

**Mechanical Properties of Mild Steels EN2C and S84 Over the Temperature Range 800° to 1200° C.** M. S. Binning and B. F. Billing. (*Royal Aircraft Establishment Techn. Note*, MET 267, 1957, July, pp. 8). Tensile, creep and scaling tests are reported.

**Studies on Mechanical Properties of Timken 16–25–6 at Elevated Temperatures. V.** T. Hasegawa and O. Ochiai. (*Tetsu-to-Hagane*, 1957, **43**, Oct., 1127–1131). Four turbine rotors in Timken 16–25–6 showed generally the same properties as those of the forged bar previously reported. When the forging effect is small in the rotor, a clear pattern of micro-segregation is seen, and the ductility is low and not uniform. The boss part is less ductile than the rim.—K. E. J.

**Equilibrium Constant of Reaction $Al_2O_3 \rightarrow 2Al + 3O$ in Liquid Iron.** H. Sawamura, T. Mori and T. Araki. (*Mem. Fac. Eng. Kyoto Univ.*, 1957, **19**, Apr., 207–219). Results previously published are confirmed and an average value is presented ($-65\,200\ T + 21{\cdot}33$).

**Diffusion in Metals.** P. Shewmon. (*Ind. Eng. Chem.*, 1958, **50**, Mar., Part 2, 492–495). A review for 1957. (88 refs.).

**Some Features of the Diffusion of Magnesium in High Resisting Iron.** I. L. Mirkin and E. P. Rikman. (*Lit. Proizv.*, 1957, (12), 13–16). The diffusion of Mg within the space of the spherical graphitic enclosures is probably irregular and occurs in strata. This supposition is indirectly confirmed by the results of the experiments described in this article and by the analysis of the methods of determining the content of Mg in the graphite. It agrees also with the laminar structure of graphite and with the fact that in combination with various elements, the layers of carbon atoms in the graphite are not destroyed.

**Brittle Fracture of Low-Carbon Steel as a Result of the Action of Hydrogen.** M. I. Chaevskii. (*Izvest. Akad. Otdel. Tekhn.*, 1957, (9), 121–122). [In Russian]. In the experiments described, low-carbon steel test pieces were subjected to tensile testing while immersed in acid liquids and undergoing galvanization or electrolytic polishing. The results showed that in galvanization, atomic hydrogen penetrates the metal with special intensity along plastic slip lines over the whole volume of the metal leading to brittle fracture (when the test pieces were covered with special bags containing $MnO_2$, embrittlement did not occur). Electrolytic polishing did not produce much effect on the tensile behaviour. The author recommends that cathodic protection should not be used for loaded parts in acid liquids.—S. K.

**Activity of Nitrogen in Liquid Pure Iron.** Y. Kasamatu and S. Matoba. (*Technology Reports of Tôhoku Univ.*, 1957, **22**, (1), 109–119; *Tetsu-to-Hagane*, 1957, **43**, Sept., 925–926). Measurements were made in a controlled $N_2$ or $N_2$-A atmosphere up to 2·3 atm. between 1560–1740° C. The equilibrium constant, free energy change and the activity coefficient were measured.

**Metallurgical Features of Structural Boron Steels.** O. N. Meshcherinova, L. I. Posysaeva and G. A. Khasin. (*Stal'*, 1958, (1), 75–81). Austenitic grain size depends on the amount of B, the more B the coarser the grain size and the greater its heterogeneity. Ductility after quenching and tempering are reduced by B, but these effects can be reduced by correct

deoxidation with Al and Ti when B is added as Fe-B. Al is added at the rate of 1 kg/t for 0·3% C steel or 0·6 kg/t for steel >0·3% C, or Ti at 0·06—0·1% depending on composition. Suitable hardenability, fine and uniform austenite grain and good mechanical properties can be attained in this way.—R. S.

**Studies on 12% Chromium Heat-Resisting Steels. VII. Effect of Boron and Boron plus Nitrogen.** T. Fujita. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1063–1065). Creep and rupture data are given for specimens containing no B or $N_2$; 0·10%B; 0·10%B and 0·02%N; and 0·03%B and 0·03%N. **VIII. Effect of B, B + N, Ti and W on 12% Chromium Steels.** (1065–1067). Creep, rupture and tempering hardness data are given for a number of steels (compositions given), including those having the additional elements stated.—K. E. J.

**Effect of Nb and N on LCN Heat-Resisting Steel. On Precipitation during Tempering after Solution Treatment. III.** T. Masumoto. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1067–1069).—K. E. J.

**Variation of the Titanium Effect with the Content of Various Elements—C, Si, Mn, P, S, N.** B. Eto. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 930–933).—K. E. J.

**Method of Determining the Mass Coefficient of Weakening of X-Rays for Powders and Solutions.** A. S. Ivoilov and L. L. Denisova. (*Zavods. Lab.*, 1958, **24**, (5), 562–565). The authors describe their method for determining mass weakening coefficients and give results of determinations characterizing its potentialities. Materials covered include magnetite, hematite and calcite. The sensitivity of the method for powders is 0·05–0·1 $cm^2.g^{-1}$ and a determination takes 15–20 min.—S. Z.

**X-Ray Structural Study of the Phase Composition of Sinter.** A. G. Slabchenko. (*Zavods. Lab.*, 1958, **24**, (5), 579–582). Results obtained from X-ray studies of sinters of basicity 0–1·1 are presented and discussed. The sinters were made in a pot, the mix consisting of 0–3 mm particles of Krivoi-Rog ore, coke and limestone. The mix contained 6·5–7% moisture and 4–8% of fuel. The changes in the diffraction patterns as a result of reduction in $H_2$ for 10, 20 and 30 min at 800° C were studied in addition to those associated with variations in the mix composition and magnetic separation. It was concluded that the FeO, $Fe_2O_3$, CaO and $SiO_2$ form a eutectic and not chemical compounds (calcium ferrites and lime-iron olivines).

**Use of Micro-X-Ray Methods for Investigating Mineral Raw Materials.** A. S. Ivoilov and D. T. Grebennikova. (*Zavods. Lab.*, 1958, **24**, (5), 582–585). The use of absorption X-ray methods for the study of magnetite, titano-magnetite and other minerals from enlarged transmission radiographs of thin sections is described. Test experiments showed that with grain sizes over 0·02 mm in sections 0·02–0·1 mm thick many minerals can be identified whose determination is difficult by microscopy in reflected or transmitted light. As an example the quantitative determination of minerals in ilmenite concentrate is described in detail. The method is as accurate as and quicker than mineralogical analysis. Typical X-radiographs are shown.

**Present State of X-Ray Structural Analysis.** K. I. Narbutt. (*Zavods. Lab.*, 1958, **24**, (5), 604–613). This survey deals with fluorescence and absorptional methods of X-ray analysis and gives an account of X-ray spectrum microanalysers.—S. K.

**Adapters for the Type URS-501 Camera for Photography at High and Low Temperatures.** L. S. Zevin and D. M. Kheiker. (*Zavods. Lab.*, 1958, **24**, (5) 636–638). The devices briefly described enable X-ray structural analysis to be carried out in the temperature range from −196° to +850° C. Specimens in the form of single crystals or polycrystalline wire or small plates are used.—S. K.

**Attachment for Studying Textures with a Type URS-501 X-Ray Unit.** M. M. Borodkina. (*Zavods. Lab.*, 1958, **24**, (5), 638–639). The attachment briefly described enables the texture of polycrystalline strip specimens to be studied by the X-ray ionization method. Results obtained for a 50%Ni–50%Fe alloy 50$\mu$ thick and 98·3% deformed are shown.

**Apparatus for the Low-Temperature X-Ray Photography of Specimens.** V. F. Mindukshev and Yu. S. Terminasov. (*Zavods. Lab.*, 1958, **24**, (5), 641). A brief account is given of an apparatus for back-reflection X-ray photography at liquid nitrogen temperature. The apparatus was used with a type KROS-1 camera.—S. K.

**X-Ray Camera for Investigating the Texture of a Thin Surface-Layer on a Flat Specimen.** B. A. Mishin. (*Zavods. Lab.*, 1958, **24**, (5), 642). A brief description is given of an X-ray camera in which there is a very small angle between the radiation beam and the surface of the flat test piece, thus enabling surface layers down to a fraction of a micron in thickness to be studied.—S. K.

**High-Temperature X-Ray Camera.** A. V. Belotskii and V. N. Gridnev. (*Zavods. Lab.*, 1958, **24**, (5), 643). In the camera briefly described it is especially easy to set the specimen and load the film. Patterns obtained from type U-12 steel at temperatures up to 850° C and with and without hardening are shown.—S. K.

**High-Temperature Camera for X-Ray Photography of Liquids and Solids.** A. S. Lashko and G. S. Svirskii. (*Zavods., Lab.*, 1958, **24**, (5), 646). The camera briefly described enables X-ray patterns from liquids or flat solids in a vacuum or inert atmospheres and at temperatures up to 900° C to be photographed.

## METALLOGRAPHY

**Summarized Proceedings of a Conference on Electron Microscopy: Bangor, Sept. 1957.** H. W. Emerton. (*Brit. J. Appl. Phys.*, 1958, **9**, Aug., 306–312). A study of creep in tempered steel by R. G. Baker and J. Nutting is summarized. Also a study of the surface of α-iron after deformation by D. G. Brandon. Thin sections of stainless steel were examined by M. J. Whelan, P. B. Hirsch, R. W. Horne and W. Bollmann; and an investigation of pure and corrosion fatigue cracks in mild steel by W. F. Jack and J. W. Sharpe, an examination of inclusions and precipitates in steels by G. R. Booker and J. Norbury; products formed in nitriding, by the same authors and A. L. Sutton; and inclusions in Al-killed steel ingots, by the same authors and A. Westrope are also briefly abstracted.

**On Grain Control of Austenite in Steel.** T. Tanoue. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 999–1000). The variation of grain-size distribution with soluble Al (0·002–0·027%) and the change of grain size with temperature are shown graphically.—K. E. J.

**On the Relation between Austenitic Grain Size and Aluminium Nitride. III. Austenitic Grain Growth and AlN in Case-hardening Steel.** K. Kato. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 1000–1002). For several specimens of carbon and Cr-Mo case-hardening steel (compositions given), the increase of austenitic grain size with heating temperature is shown graphically, together with the AlN content at each temperature. The results of microscopical analysis of the AlN are given.

**The Relation Between the Cold Working, Annealing and Ferrite Grain-Size of Electric Resistance-Welded Pipes. II. Rimmed Steel Pipes.** T. Yamamoto. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 915–917). Data are given for the hardness of the matrix for three steel samples given reductions between 6% and 48% and annealing treatments of varying length between 600° and 750° C, and also for the ferrite grain size of the rim and core of the tubes for degrees of reduction up to 40%.

**Structure Abnormalities of Sulphur Iron.** A. A. Zhukov. (*Lit. Proizv.*, 1957, (12), 16–20). It is the silicon and not the graphite, which plays the greater rôle in the appearance of the extra centres of graphitization in iron. The tendency of low-carbon iron to whitening (formation of cementite) is explained by the accumulation of Si in the developing dendritic skeleton of the melt, and by the impoverishment of the Si in the interdendritic space. This causes a disturbance in the separation of graphite, and hence increases the tendency to the formation of ledeburite.—L. H.

**Recrystallisation and Spheroidisation of the Cementite in Steel 35 After Cold-Reduction.** S. Orzechowski. (*Prace Inst. Ministerstwa Hutnictwa*, 1957, **9**, (5), 191–202). [In Polish]. The influence of temperature and time on the processes which take place during annealing, and the properties of sheets of steel (Mark 35) have been investigated. It was found that the rate of heating has a definite effect. The effect of the primary structure was also examined and the rate and temperature of annealing and the spheroidization process on cementite investigated.

**Metasomatism in Metallurgy.** R. Skorski. (*Prace Inst. Mech.*, 1957, **6**, (20), 65–68). ]In Polish]. The author explains the term metasomatism in metallurgy and describes two such phenomena: the deposition of Ti on iron during heat-treatment of the latter in a molten mixture of $TiO_2$ with $Na_2CO_3$, and the change of Mn into a Bi-Mn alloy, which takes place when a piece of Mn is immersed in molten Bi and is caused by rapid diffusion (8 mm per h at 800° C).

**Graphitization of Pearlitic Cementite in Spheroidal Cast Irons at Hypoeutectoid Temperatures.** S. Drápal. (*Hutn. Listy*, 1958, **13**, (1), 42–49). [In Czech]. Two series of cast irons were used, one with 2·18% Si, the other with 2·76%. Within each series the Cr contents were 0·04%, 0·21%, 0·47% or 0·75%. It was found that (*a*) the pronounced graphitization of pearlitic cementite at hypoeutectoid temperatures occurs only after a definite time interval, during which spheroidization and coagulation of cementite takes place. (*b*) The length of this incubation period increases with increasing Cr content, decreasing temperature and decreasing Si content. (*c*) The graphitization reaction can be represented in its time-dependence by the relation proposed by Hickley and Wood (*J.I.S.I.*, 1954, **176**, 129–139). (*d*) In the process of graphitization of the pearlitic cementite the Cr content of the remaining complex cementite increases, thus stabilising the cementite and facilitating the retention of some undecomposed complex cementite in the cast iron. (*e*) The amount of retained cementite increases with the Cr content of the iron; (*f*) the amount decreases with increasing Si content, the latter contributing to the instability of the cementite. (*g*) The time required for the completion of the graphitization was found to obey the empirical relation $t = Aexp(B/T)$, where T is the temperature in °K and A and B are constants.—P. F.

**The Pearlite Transformation in Alloy Steels.** M. E. Blanter. (*Metallov. Obra. Met.*, 1955, (4), 1–15). The basis of the accepted hypothesis for the transformation is the preliminary diffusion of the alloy elements in the austenite for the formation of concentrations suitable for the formation of stable carbide and ferrite. This hypothesis, however, has no sufficient theoretical and experimental confirmation. The author refers to his work since 1948. The alloy elements determine the kinetics of the pearlite transformation as a result of their action on the three determining factors: the diffusion of carbon in austenite, the inhibition of cold hardening of austenite and the position of the critical temperature point $A_1$.—L. H.

**Studies on Duplex-Grain Structures of Austenite. III. Dependence of Duplex Grains on Segregation.** Y. Masuko. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 973–975). The effect of homogenization on the degree of duplex-grain formation and segregation is investigated in Ni-Cr-Mo case-hardening steels (analyses given), and various carbon steels are also discussed.—K. E. J.

**Behaviour of Residual Austenite in Isothermally Quenched Grey Cast Iron, During the Process of Wear.** A. N. Chukhleb. (*Metallov. Obra. Met.*, 1956, (12), 33–35). When operating under abrasive conditions, the high wear resistance of grey cast iron is due to the γ–α phase transformation which occurs at the frictional surface in a very thin layer (0·001 mm) as shown by X-ray examination.

**Austenitising Behaviour and Rapid Life Test of Drawn Wire for Bearing Balls.** M. Ueno and Y. Nakano. (*Tetsu-to-Hagane*,

1957, **43**, Oct., 1137–1142). Physical properties, austenitising behaviour and rapid life-test data are given for five steels (compositions given), and the effects of ingot size, raw materials and V additions are followed.

**Isothermal Transformation of Austenite in High Speed Steels.** A. P. Gulyaev and K. A. Malinina. (*Metallov. Obra. Met.*, 1956, (12), 2–8) The results of the investigations of the kinetics of the breakdown of austenite in high speed steel are mainly used in establishing correct heat treatments for tool steels. It is necessary to differentiate between: (*a*) the austenite obtained after heating to a temperature above the critical point and then cooling to temperatures below $A_1$ but above $M_s$ (the primary austenite); (*b*) the residual austenite retained after cooling below $M_s$ and present along with martensite in the structure of the quenched steel; (*c*) the austenite which exists along with martensite in the structure of the tempered steel. Some heat treatments are laid down.—R. S.

**The Kinetics of the Formation of Bainite in High-purity Iron–Carbon Alloys.** S. V. Radcliffe and E. C. Rollason. (*JISI*, 1959, **191**, Jan., 45–54). [This issue].

**The Isothermal Transformation of Austenite in Carbon Steels Containing 0·5% C and 0·18% C. An Electron Microscopical Investigation.** S. Modin. *Jernkont. Ann.*, 1958, **142**, (2), 37–80). Pearlite and bainite formation has been examined partly by replicas and partly by isolation of carbides. Many micrographs are shown and discussed.

**The Effect of Cooling Rate from the Austenite Range upon the Microstructure of Low Carbon Rimmed Steel. On the Formation of a Pearlite-free Structure.** K. Yoshida. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 980–982).

**Transformation of Residual Austenite: the Basic Cause of Deformation of Welded Structures.** A. V. Mordvintseva. (*Metallov. Obra. Met.*, 1956, (7), 48–54). It is shown that the transformation of residual austenite at room temperature is the fundamental cause of the deformation of low-alloy steel welded structures. Former opinions to the effect that no deformation occurs in welded structures after cooling are valid only in relation to a limited number of steels.—R. S.

**Effect of Residual Austenite Obtained During Quenching on the Impact Strength and Form of Fracture in Highly Tempered CrNiMo Steel.** A. P. Gulyaev and I. I. Slavina. (*Metallov. Obra. Met.*, 1956, (3), 15–17). Although, after fully tempering, residual austenite is completely transformed, it was thought to be of interest to clarify the question as to whether the amount of retained austenite in quenched steel does affect the form of fracture and the impact strength. For this purpose, a steel containing C 0·26, Cr 1·3, Ni 3·9, Mo 0·29 was heated at various rates to temperatures in the range of 800 to 1000° C and quenched at different speeds in water and oil or was air- or furnace-cooled. Results obtained in each case are described.—R. S.

**Phase Transformation in Austenitic Steels Under the Influence of Stresses.** A. V. Stepanov. (*Metallov. Obra. Met.*, 1955, (5), 42–52). A Review. Analysis has shown a new phase with a hexagonal close-packed crystal lattice during the decay of the $\gamma$-phase into $\alpha_2$. It is analogous to phase $\epsilon$ in the system Fe-Mn. The cold deformation corresponds to the transformation of the $\gamma$-phase in Fe-Cr-Ni alloys to $\alpha_2$.—L. H.

**The Effect of Vibrations, Especially Magnetostrictive, on the Transformation of Residual Austenite into Martensite.** E. Zmiharski. (*Prace Inst. Mech.*, 1957, **6**, (19), 30–33). [In Polish]. The author investigated the effect of vibrations on the change of properties and the microstructure of hardened steel. The effect of magnetostrictive vibrations is so considerable that it exceeds in efficacy the tempering (at 210° C) of hardened steel.

**Relation between the Position of the Martensite Point and the Austenite State.** G. A. Oding. (*Metallov. Obra. Met.*, 1955, (4), 28–31). The rate of the incipient growth of martensite needles soon establishes the form of the martensite crystal, a form of lamina. The clearly audible crackle accompanying the martensite reaction in iron-nickel alloy, may be explained by the fact that the movements preceding the appearance of the martensite crystals in the austenite represent the gradual accumulation of fields of stress of the second order, which lead to the development of tangential stresses of measurable magnitudes. Consequently the thermodynamic instability of the $\gamma$-lattice in the more stable $\alpha$-lattice may mean the beginning of martensite crystal formation.—L. H.

**The Problem of Tchernoff's *b* Point.** C. Niewiadomski. (*Prace Inst. Mech.*, 1957, **6**, (19), 34–38). [In Polish]. The latest research by W. D. Sadowski (Russia) confirms the existence of the *b* point discovered by Tchernoff. The author gives a detailed review of investigations conducted in this field. The recrystalization of austenite is discussed, and further research is recommended.

**The Microstructure of Cast High-Speed Steels.** P. M. Yushkevich and S. A. Fedorova. (*Metallov. Obra. Met.*, 1955, (3), 21–25). At room temperature in forging, high-speed steels present their own heterogeneous system of austenite, martensite, ferrite, carbides and tungstides. Of particular interest is the separation of ledeburite within the structure of forged high-speed steel on heating to 1320–1380°, i.e., in the range between the eutectic and the beginning of crystallization of the $\gamma$ phase. Ledeburite was obtained experimentally on pouring the melt to form bars in earthen moulds of 12 mm diameter with subsequent tempering from 1320–1380°, when the metal is a viscous mass. These experiments point to the possibility of obtaining a structure from these cast steels without the formation of the ledeburite lattice.—L. H.

**Effect of Titanium Deoxidation on the Mechanical Properties and Microstructure of Cast Cr-Mo-V Steel.** R. D. Ahles, C. H. Wright and R. S. Zeno. (*Trans. A.F.S.*, 1956, **64**, 591–599). Mechanical properties, type of inclusion and structure were examined as a function of S content. Ti was added from 0–2 lb/ton with 0·01–0·028% S. Tensile ductility and quality factor decrease with rise of S or Ti. Cast steel deoxidised with Ti (2 lb/t) resisted tempering. With 0·017% S max. deoxidised with Ti up to 1 lb/t, good tensile properties were obtained. Localised sulphide inclusions were always observed but not in sufficient quantity to affect ductility.

**Effect of Aluminium Deoxidation on the Mechanical Properties and Microstructure of Cast Cr-Mo-V Steel.** R. S. Zeno, R. D. Ahles and W. R. Nestle. (*Trans. A.F.S.*, 1956, **64**, 600–610). Inclusions and microstructure were also examined. Additions of Al from 0–2 lb/t at S contents of 0·011–0·027% were used. It was concluded that with 0·011–0·018% S, additions of Al caused little loss of tensile ductility, with Type 1 sulphide inclusions. For 0·024–0·027% S an appreciable ductility loss resulted and Type II inclusions tended to form. No change of structure was observed with the former S range and decrease of mechanical properties suggested an intrinsic effect of Al. Some observations on the effects of oxygen were made.

**Possible Existence of $Fe_3Ni$.** Y. Tino. (*Sci. Rep. Tôhoku Univ.*, 1956, **40**, Aug., 17–23). Studies of $\gamma$-phase alloys with about 30% Ni are reported.

**An Investigation on the Titanium-Iron-Carbon System.** Y. Murakami, H. Kimura and Y. Nishimura. (*Mem. Fac. Eng. Kyoto Univ.*, 1957, **19**, July, 302–319). The solubility ranges of the TiFe and $TiFe_2$ phases, properties of the $\delta$-phase and its relationships with $\alpha$-iron and $TiFe_2$ and the invariant reactions in the various phases are determined.

## CORROSION

**How Oxidative Corrosion Works.** R. V. Jelinek. (*Chem. Eng.*, 1958, **65**, Aug. 25, 125–130). An account of fundamental principles with notes on bimetallic effects.

**Accelerated Tests on the Anticorrosive Resistance of Chemically Passivated Electrolytic Zinc Deposits on Steel.** T. Biestek. (*Prace Inst. Mech.*, 1957, **6**, (19), 39–54). [In Polish]. The purpose of Zn coating passivation, the selection of a method, the composition of the bath, the structure and properties of passive layers, the effect of passive protective layers and methods of accelerated corrosive testing are discussed. Extensive researches conducted by the author are described and a critical survey of the results is given.

**Salt Spray Corrosion Tests.** N. F. Murphy. (*Products Fin.*, 1958, **22**, Mar., 56, 58–59). A brief comparison with exposure test findings is given and the possibility of correlation is suggested.

**Methods of Estimating the Corrosion Resistance of Electrolytic Zinc Coatings on Steel.** T. Biestek. (*Prace Inst. Mech.*, 1956, **5**, (16), 1–18). [In Polish]. Application of Zn plating and its protective character is discussed. A comparison between Zn and Cd coatings is given. Investigations of porosity, uniformity of coating and corrosion-resistance of Zn coatings was carried out. The author gives a detailed account of his own investigations of corrosion-resistance of electrolytic Zn coatings. Accelerated corrosion experiments were conducted by dipping in a 10% solution of $NH_3Cl$, in a moist atmosphere at high temperatures with Gardner's apparatus, in a salt bath and in hot distilled water. Results are illustrated and discussed.

**Methods of Testing Rust-Resisting Steels in Chemical Solutions.** H. Zitter. (*Werks. Korr.*, 1957, **8**, Dec., 746–760). The author discusses, in the light of the literature and of original investigations, the laboratory testing of rust-resisting steels, with special reference to assessing their service behaviour. He divides the factors concerned into three groups: (i) the heat-treatment and surface preparation of the material; (ii) the corrosive medium; (iii) the type of test and the method of conducting it. Of these the first are of particular importance. A simple testing procedure for resistance to mineral acids is described and some results obtained by it are given.—J. C. H.

**Photogalvanic Investigations of the Passivation of Iron in Nitric Acid.** I. B. Rabinovich, V. I. Veselovskii and L. N. Bokov. (*Zhur. Fiz. Khim.*, 1955, **29**, (12), 2218–2221). [In Russian]. It is shown that in concentrated $HNO_3$ iron shows high photogalvanic activity, and that passivation proceeds in much the same way as when an oxide layer is formed during heating.—L. H.

**Inhibition by the Surface Layers of Some Metals, of the Electrodiffusion of Hydrogen in Iron and Steel.** L. I. Freiman and V. A. Titov. (*Zhur. Fiz. Khim.*, 1956, **30**, (4), 882–888). [In Russian]. An explanation is offered of the mechanism of "poisoning" of the surface of iron by As in acid solutions which arise from corrosion processes. Under the conditions of the experiments described, thin layers of tin and lead speed up electrodiffusion of $H_2$, but sufficiently thick layers of these metals inhibit it.—L. H.

**Industry Fights Corrosion.** (*Proceedings of the Corrosion Convention and Corrosion Techn.* 1957. London. 1958, pp. 107). **Protection of Plant and Equipment in the Petroleum and Chemical Industries.** H. B. Footner. (1–5; discussion 5–7). A review of working environments, painting and related treatments of outside steelwork, tanks, pipelines and drums. **Corrosion in the Shipping Industry.** S. R. Bew. (8–11; discussion 12–13). A brief review of locations, protective methods and costs. **Packaging to Prevent Corrosion in Transit and Storage.** F. A. Paine and D. Watkinson. (14–17; discussion 18–19). Causes of corrosion and selection and testing of packaging materials are surveyed. **Corrosion and Water Treatment.** G. G. Sindery. (20–33; discussion 33–34). Mechanisms of corrosion and the effects of pH and dissolved $O_2$ are outlined, the effects of hydrolisable salts is explained and treatment methods are reviewed. Boiler plant, steam heated appliances, cooling systems and hot and cold water lines are then briefly considered. Protective measures and the use of tannins as inhibitors

are included. **Corrosion in the Atomic Energy Industry.** M. D. Jepson, D. White and B. L. Harbourne. (35–39; discussion 39–40). A general account, mainly of the uses of non-ferrous metals. **Fuel Additives in the Fight Against Corrosion.** B. J. Zaczek and R. Grindley. (41–44; discussion 44–47). Sulphur corrosion and the effects of additives and their evaluation are reviewed. **Paints and the Part They Play in the Fight Against Corrosion.** H. Hollis and L. J. Coleman. (48–53; discussion 54–55). Primers are briefly reviewed as are undercoats and finishing coats and causes of failure are listed and typical effects are shown. **Selection of Corrosion-Resistant Metals and Alloys.** J. B. Cotton. (56–62, discussion 62–63). Mainly non-ferrous, though steels resistant to sulphuric acid are mentioned. **Hot-Dip Galvanizing.** W. L. Hall. (72–75, discussion 76–77). The process, the nature and behaviour of the coatings are described. **Some Aspects of Cathodic Protection.** W. G. Waite. (78–88, discussion 88–90). A general review of the uses of Zn, Mg and its alloys and Al and of the impressed current methods. Uses are surveyed and economic advantages assessed. **Sprayed Metal and Plastics Coatings.** G. H. Jenner. (91–96, discussion 96–97). **Anti-Corrosion Coatings for Buried Pipes.** W. D. Parker and A. G. Wilkie. (98–105; discussion 106–107).

**Corrosion of Steel: I. Protective Action of Oxides of Lead and Zinc.** A. K. Choudhury and S. C. Shome. (*J. Sci. Ind. Res.*, 1958, **17A**, Jan., 30–34). A study on underwater corrosion is reported, mixtures of red lead with litharge or zinc oxide protected steel specimens more effectively than red lead alone. Litharge appears to improve the physical structure of the paint and zinc oxide to have also an inhibitive action. **II. Inhibitive Behaviour of Chromates, Molybdates and Tungstates of Zinc and Lead.** (35–38). Aqueous extracts of the zinc salts inhibit corrosion, the molybdate being most effective, chromate produces waterline attack. The tungstate is less effective. Lead salts have no protective action being insoluble. In oil paints molybdate retains and chromate loses its corrosion inhibiting action. The effects do not correlate with alkalinity.

**Mechanism of Internal Corrosion of Water Pipes.** S. Jankovic. (*Zastita Mat.*, 1956, **4**, Dec., 400–402).

**Passivation and the Formation of Protective Layers in Water.** L. Piatti. (*Z. Metallk.*, 1958, **49**, Jan., 42–47). The corrosion of metals submerged in water may be prevented by formation of a dense, adhering and non-porous coating resulting from contact of the metal with water. A basic carbonate film forms on ferrous metals. Additions of phosphates, silicates and chromates are often desirable. Sodium nitrate favours the formation of oxide films on the metal. Organic inhibitors are also effective.—R. P.

**The Influence of Surface Treatment of Metals and Alloys on their Resistance to Corrosion.** S. Mrowec and T. Werber. (*Hutnik*, 1957, **24**, April, 142–147). [In Polish]. The results of recent work show a clear influence of different kinds of surface treatment of metal and alloys on the rate of oxidation of their surfaces. The results of investigations in this field are tabulated and different theoretical interpretations are given. The theoretical interpretation of this problem is very difficult in the case of alloys because of its complexity and the lack of experimental data.

**An Unusual Corrosion Failure in Steam Superheater Tubes.** C. Phillips jun. (*Corrosion*, 1958, **14**, Feb., 57–58). Leaks occurred due to internal pitting at points in superheater tubes under external clamps. Pits were elongated along the tube and not associated with cracks. The tubes retained good mechanical strength and the steam was not particularly corrosive. Pitting is probably due to breaks in the oxide scale inside the tubes due to thermal contraction of the clamps when the outsides were washed while steam filled to remove external sodium sulphate scale.

**Corrosion Phenomena in Condenser Tubes.** A. J. Maurin. (*Corros. et Anticorros.*, 1957, **5**, Oct., 275–285). The function of condensers and their materials are discussed from the point of view of corrosion. Some examples of attack by corrosion are given from several countries. Means of corrosion prevention and control are recommended, the most important being the formation of a protective film during the first days of functioning.—R. P.

**A Contribution to the Study of the Resistance to Corrosion of Nodular Cast Iron.** I. Cornea. (*Rev. Mét. Bucarest*, 1956, **1**, 63–71). [In French]. Nodular and grey cast irons were compared for their resistance to HCl. It was supposed that, in oxygen-free acid media, the corrosion would be related to the ratio of cathodic to anodic areas and as graphite is the principal cathode, the nodular irons should be more resistant. In general this was the case, the resistance being double that of a grey iron of similar composition and microstructure.

**Iron-Silicate Slag Network Helps Wrought Iron Resist Corrosion.** E. P. Best. (*Corrosion*, 1958, **14**, Feb., *Technical Topics*, 118–120). The principal characteristics of wrought iron and its method of manufacture are outlined. Reasons for corrosion resistance are given and the tensile properties listed.—J. F. S.

**Corrosion of Metals in Tropical Environments. Part 1—Five Non-Ferrous Metals and a Structural Steel.** B. W. Forgeson, C. R. Southwell, A. L. Alexander, H. W. Mundt and L. J. Thompson. (*Corrosion*, 1958, **14**, Feb., 73–81). Corrosion tests of eight years' duration are reported for sea-water (half-tide and continuous) immersion, freshwater immersion, and marine and inland tropical atmospheric exposures in the Panama Canal Zone. Weight losses, pit-depths and tensile strength changes are included for Al, Pb, Ni, Cu, Zn and an 0·2% C low Cu structural steel. Corrosion rates are, in general, considerably higher than in comparable environments in temperate latitudes.—J. F. S.

**The Corrosion of Metals and of Metallic Protective Coatings in Industrial Atmospheres.** G. Schikorr. (*Schweiz. Arch. Wiss. Techn.*, 1958, **24**, Feb., 33–46). The author discusses the composition of industrial atmospheres in various areas in relation to corrosion of ferrous and non-ferrous metals. The $SO_2$ content of air is the main corrosive agent, with water and oxygen next. The corrosion product of iron and $SO_2$ is unstable, oxidation and hydrolysis resulting in formation of sulphuric acid. From oxide acts as an inhibitor but many corrosion processes need to be clarified further. (78 references).—R. P.

**Corrosion of Steel in Steel-Works Atmospheres.** A. V. Shreider and A. I. Sokolikov. (*Zhur. Priklad. Khim.*, 1957, **30**, (11), 1706–1708). An account is given of the investigation at the "Serp i Molot" works of the corrosion of specimens of Armco (0·033%C, 0·05%Mn, 0·027%S, 0·011%P, 0·013%Ni, 0·18%Cu) and the following steels: 30KhGSA (0·30%C, 0·95%Mn, 0·95%Si, 0·033%S, 0·030%P, 0·83%Cr, 0·25%Ni); St. 40 (0·39%C, 0·56%Mn, 0·25%Si, 0·038%S, 0·040%P, 0·28%Cr, 0·23%Ni) and NL-2 (0·18%C, 0·61%Mn, 0·33%Si, 0·025%S, 0·017%P, 0·67%Cr, 0·46%Ni, 0·46%Cu). Specimens were exposed in the pickling shop, O.H.-shop and on the works roof. Corrosion curves for each are given and the aggressiveness of the atmospheres were compared with the aid of a special cell. Among the conclusions is one that the copper-bearing steels are specially corrosion resistant only in weakly aggressive media.—S. K.

**The Corrosive Properties of Artificially Produced Layers of Hard Zinc.** W. Rädeker. (*Metalloberfläche*, 1958, **12**, Apr., 102–104). Experiments confirmed the previously observed improvement of the resistance to corrosion of zinc coatings by annealing above 530° C. Articles with a matt finish due to longer immersion in the bath were shown to possess at least as great a resistance to corrosion as those plated with a bright finish. The low deformability and less attractive appearance of the annealed material limits its use.—L. D. H.

**Damage to Solids by Liquid Impact at Supersonic Speeds.** F. P. Bowden and J. H. Brunton. (*Nature*, 1958, **181**, Mar. 29, 873–875). The effects of impact on stainless steel and on other metals and non-metals are shown.

**The Nature of Pitting Attack on Cast Iron Propellers and Preliminary Experiments on Its Prevention by Cathodic Protection.** R. I. Higgins. (*B.C.I.R.A. J. Res. Dev.*, 1957, **7**, Dec., 129–143). The nature of this type of pitting attack is discussed and evidence is presented to show that the metal is removed by a corrosion rather than by an erosion process. It is shown that the pitting attack can be prevented by cathodic protection using sacrificial Mg anodes.—B. G. B.

**Contribution to the Study of Erosion Due to Abrasion in Hydraulic Turbines.** T. Bovet. (*Schweiz. Arch. Wiss. Techn.*, 1957, **23**, Sept., 377–388). This paper reviews the mechanism of erosion in hydraulic turbines and suggests preventive measures. Suspended particles should be removed from the feed water and the material used for turbine construction should be suitably selected according to experience and local conditions.—R. P.

**Corrosion in Amine Gas Treating Solutions.** S. Lang and J. F. Mason jun. (*Corrosion*, 1958, **14**, Feb., 105–108). Laboratory corrosion tests were conducted on steel, stainless steel, Monel and nickel in amine solutions saturated with $CO_2$ and $H_2S$. The effects of temperature, pressure, amine concentration and velocity and the presence of sulphide sludge were studied. Mild steel is satisfactory in pure amine only. Otherwise stainless steel or nickel alloys are desirable, particularly if $CO_2$ is present. Field tests are described and the results interpreted in terms of operating practice as well as choice of metal.—J. F. S.

**High Temperature Hydrogen Sulphide Corrosion in Commercial Sovaformer Units. Part 2.** E. B. Backensto, R. D. Drew and J. N. Vlachos. (*Corros. Techn.*, 1958, **5**, Feb., 53–56). The authors present and discuss test results obtained in two commercial sovaformers and show that they agree with laboratory corrosion tests in $H_2S$ atmospheres. The data show that both temperature and $H_2S$ concentration are basic variables in the corrosion. Aluminizing was found to provide protection from sulphide scaling attack and there was evidence that intergranular microfissuring at the surface of low-Cr steels may occur in commercial reforming installations.—L. E. W.

**The Scaling of Pure Iron in Oxygen as a Function of Temperature.** N. G. Schmahl, H. Baumann and H. Schenck. (*Arch. Eisenhütt.*, 1958, **29**, Feb., 83–88). Measurements between 480–1080° C have shown that at the wüstite point (566°), at the Curie point of the magnetite (579°), at 616°, at the Curie point of the iron (~ 760°), and probably at 910°, the curves should be subdivided into practically rectilinear sections.—L. J. L.

**The Rate of Scaling of Pure Iron in Oxygen Under Reduced Pressures and in Carbon Dioxide.** N. G. Schmahl, H. Baumann and H. Schenck. (*Arch. Eisenhütt.*, 1958, **29**, Mar., 147–152). The rate of scaling of electrolytic iron was investigated at about 950° C in the pressure range of 2 to 760 mm. Between 76 and 760 mm, scaling was related to pressure by a parabolic law, whereas at 2 mm the relationship was linear, as already established by Pfeiffer and Laubmeyer (*Z. Electrochem., Ber. Bunsenges Phys. Chem.*, 1955, **59**, 579/583). The scaling of pure iron in a current of $CO_2$ was also investigated, the CO content produced being derived from the increase in weight of the samples.

**A Theoretical Analysis of the Excessively High Temperatures Occurring During the Scaling of Pure Iron.** N. G. Schmahl, H. Baumann and H. Schenck. (*Arch. Eisenhütt.*, 1958, **29**, Jan., 41–46). The occurrence of temperature peaks during scaling having been demonstrated in a previous study, a mathematical description of their development is now presented. Direct measurement of temperatures in samples of 1·5 to 5 mm thickness with varying specific surface has shown that thin samples exhibit higher excess temperatures during scaling at 950° C in pure oxygen than do thick samples. A simplified equation for the quasi-stationary state during scaling is presented.—L. J. L.

**Gas Corrosion of Austenitic Steels at High Temperature.** E. A. Davidovskaya and L. P. Kestel'. (*Metall. Obra. Met.*, 1956, (10), 29–34). A group of steels (EI 417; 437; 673; 395; 434; 405 and LA 1) showed little oxidation in air at 650°. (0·003–0·017 mm/year). Other steels, but in a medium of gas turbine atmosphere composition showed a more pronounced oxidation 0·003–0·098 mm/year. Addition of 0·3% of sulphur in gas for the fuel mixtures accelerates oxidation, which reaches in these conditions 0·005–0·17 mm/year.—L. H.

**A Comparison of Methods of Marking in Investigations of the Mechanism of Oxidation of Metals and Alloys.** L. Czerski, S. Mrowec and T. Werber. (*Arch. Hutn.*, 1958, **3**, (1), 37–48). The layer, microscopical and autoradiographic methods of detecting markers in the scale are compared. The second method is recommended. A formula is derived for calculation of weight fraction of oxidizing agent in the inward diffusion process from the position of the marker. (21 references).

**Wüstite Decay in the Presence of Metallic Iron after Quenching to Temperatures below 570° C.** W. A. Fischer and A. Hoffmann. (*Arch. Eisenhütt.*, 1958, **29**, Feb., 107–113). Wüstite samples were melted in quartz tubes at a pressure of about $10^{-4}$ mm, and annealed for 1 h at 1100° or 640° C, quenched to temperatures between 145° and 550° C, and then maintained at constant temperature for periods of up to 30 days. After quenching in ice-water, wüstite decay was studied under the microscope, and by X-ray photography.

**Influence of Insulation Design on Condensation and Corrosion.** N. L. Dolbey. (*Corros. Techn.*, 1958, **5**, Feb., 41–44). The author first discusses briefly typical temperature gradients through insulated and uninsulated structures with impermeable outer shells and shows how these gradients determine the amount and location of condensation. The use of water-repellent vapour-permeable insulation is criticized since water can condense in it. The author also points out that an air space between the insulation and the metal surface can lead to greater condensation than if no insulation were applied. The importance of properly designed insulation and the selection of suitable materials is stressed. By selecting an insulation with good capillary properties and applying this directly to the metal surface, any water condensing on the metal is withdrawn in the body of the insulation and passes to the opposite side where, if there is an air circulation over the warm face of the insulation, the water is re-evaporated. The author concludes by discussing briefly the various faulty insulation procedures in ships, railway carriages, etc.

## ANALYSIS

**Preparation and Examination of Chemical Analysis Control Samples.** A. Stetter. (*Stahl Eisen*, 1958, **78**, Aug. 7, 1126–1129). Directions for the preparation and examination of chemical analysis control samples are given. The procedures to be followed for ensuring uniformity of the samples and accuracy of the examination include sample taking, processing and preparation of the raw materials as well as chemical analysis. A number of examples is quoted.—T. G.

**Analytical Use of the Reaction of Ferrous Salts with Violuric Acid.** P. Černy. (*Coll. Czechoslovak Chem. Comm.*, 1958, **23**, Jan., 105–109). [In German]. Photometric determination is described and the effects of other ions are tabulated.

**Quinoxaline-2 : 3-dithiol as a Colorimetric Reagent. Determination of Nickel in Ammoniacal Solutions.** D. A. Skoog, Ming-Gon Lai and A. Furst. (*Anal. Chem.*, 1958, **30**, Mar., 365–368).

**Nitroso-R-Salt. Reagent for Cobalt and Iron.** R. J. Reynolds and P. Casapieri. (*Organic Reagents for Metals, Monograph* No. 33, 1957, pp. 7).

**Amperometric Determination of Cobalt in Steel.** K. Szmidt and J. Weber. (*Prace Inst. Mech.*, 1957, **6**, (19), 71–75). [In Polish]. Methods of determination of Co in steel are reviewed. The possibilities of determining Co in steel by means of amperometric titration have been investigated. High-speed steels containing Cr, Mn, V, Mo and W have been analysed for Co by this and by classical methods. The results did not differ by more than 1·5% of the total Co content.

**Coulometric Determination of Chromium.** L. Meites. (*Anal. Chim. Acta*, 1958, **18**, April, 364–372). A method eliminating Cu, Pb, V and other elements and interfered with only by V and large amounts of Mo is described.

**Microdetermination of Chromium with 1 : 5-Diphenylcarbonhydrazide.** T. L. Allen. (*Anal. Chem.*, 1958, **30**, Mar., 447–450).

**Photocolorimetric Determination of Zirconium by Means of Morin.** H. Tůma and N. Tietz. (*Coll. Czechoslovak Chem. Comm.*, 1958, **23**, Jan., 142–146). [In German].

**Rapid Determination of Small Amounts of Lead in Electrolytic Manganese and Ferromanganese.** H. Goto and K. Hirokawa. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1958, **10A**, Feb., 11–19). [In English]. Dithizone is used.

**Microdetermination of Zirconium in Sulphuric Acid Solutions with Pyrocatechol Violet.** J. P. Young, J. R. French and J. C. White. (*Anal. Chem.*, 1958, **30**, Mar., 422–425).

**Coulometric Reduction of Oxides on Tin Plate.** R. P. Frankenthal, T. J. Butler and R. T. Davis, jun. (*Anal. Chem.*, 1958, **30**, Mar., 441–443).

**Polarographic Determination of Copper and Bismuth in Minerals and in Iron.** J. Doležal and J. Novák. (*Chem. Listy*, 1958, **52**, (1), 36–39). [In Czech]. A rapid polarographic method is described. About 0·01% of Cu in iron can be determined with adequate accuracy.—P. F.

**Study on the Chemical Analysis of Sand Iron and Ilmenite.** S. Wakamatsu. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 886–887). [In Japanese].—K. E. J.

**Determination of Alumina, Lime and Magnesia in Manganese Ores, by Mercury Cathode Electrolysis.** Y. Yoshino. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 888–889). [In Japanese].—K. E. J.

**Determination of Sulphur in Basic Slag by the Combustion Method. I. High-frequency Combustion—EDTA Titration Method.** Y. Yoneyama. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 889–891).—K. E. J.

**Rapid Analysis of Some Silicate Materials.** H. Bennett, W. G. Hawley and R. P. Eardley. (*Trans. Brit. Ceram. Soc.*, 1958, **57**, Jan., 1–28). Analysis of clay and fired products made up of the usual oxides is described.

**Analytical Methods for Trace Quantities of Cyanide Compounds in the Plating Waste Effluent.** W. F. Stevens and L. E. Lancy. (*Proc. Ann. Conv. Amer. Electroplaters' Soc.*, 1957, 70–72; discussion, 162).

**Recovery of Methyl Acetate from the Waste Solution Used for Determining the Aluminium Nitride in Steel.** S. Yamasaki. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 953–954).—K. E. J.

**The Direct Determination of Mineral Matter in Coal.** M. Bishop and D. L. Ward. (*Fuel*, 1958, **37**, Apr., 191–200). The method of Radmacher and Mohrhauer is slightly modified.

**Automatic Unit for Determination of Volatile Matter in Coal, Coke and Char.** R. P. Hensel and S. A. Jones. (*Anal. Chem.*, 1958, **30**, Mar., 402–404).

**Determination of Nitrogen in Coke by the Semimicro Kjeldahl Method.** R. A. Mott and H. C. Wilkinson. (*Fuel*, 1958, **37**, Apr., 151–158).

**Determination of Nitrogen in Coke by the Dumas-Schöniger Method.** R. Belcher and M. K. Bhatty. (*Fuel*, 1958, **37**, Apr., 159–165).

**Modern Spectrographic Analysis of Steel Works Materials.** D. Manterfield. (*Instr. Practice*, 1958, **12**, Apr., 359–366). Work in the visible and for ultra-violet, instruments and procedures are outlined.

**A Triggered 220-V a.c. Spark for High Precision in Metal Analysis.** C. G. Carlsson. (*Colloquium Spectroscopicum Internat. VI. Amsterdam* 1956, reprint, 370–376). The circuit is given and analyses of cast iron and steel are given as examples. The circuit parameters are of great importance and this is demonstrated by oscilloscope patterns.

**Spectrophotometric Determination of Iron with Ethylenediaminedi(o-hydroxyphenylacetic acid).** A. L. Underwood. (*Anal. Chem.*, 1958, **30**, Jan., 44–47).

**Spectrographic Simultaneous Determination of Minor Elements (Ni, Cr, Sn, As, Al) in Carbon Steel Using Plain Samples.** M. Ihida. (*Tetsu-to-Hagane*, 1957, **43**, Sept., 892–894).—K. E. J.

**Spectrochemical Determination of Lead in Steel.** L. C. Flickinger, E. W. Polley and F. A. Galletta. (*Anal. Chem.*, 1957, **29**, Dec., 1778–1779). Direct determination of 0·05–0·5% Pb in a cupped electrode using synthetic standards is described.

**The Fluorite Polychromator for Determining Carbon, Phosphorus and Sulphur in Steel.** (*Wild Barfield J.*, 1958, **6**, Mar., 2–6, *Hilger Journal*, **4**, (2)).

## INDUSTRIAL USES AND APPLICATIONS

**Austenitic Steel Castings for the Construction of Gas and Steam Turbines.** W. Stauffer and A. Keller. (*Giesserei*, 1958, **45**, Apr. 24, 229–239). The various types of austenitic steel and their application in this field are first described. The creep strengths of various types are given in graph form. The notch toughness, embrittlement and the welding of austenitic cast steels are discussed and illustrated with numerous micrographs. The problems incurred and the specifications for the development of austenitic creep-resisting steel castings are enumerated.—R. J. W.

**The Development of Special Steels for the Electrical Industry.** P. Jehlička and R. Vokáč. (*Výzkumné Práce z Oboru Oceli a Železa, State Publ. Ho. Tech. Lit., Prague*, 1958, 189–196). [In Czech]. Details are given of the development of a low-carbon low-alloy steel suitable for the manufacture of grids and anodes in valves where nickel is not essential. The best steel contained 0·025%C, 0·22%Mn, 0·7%Si, 0·013%P, 0·017%S, 0·11%Cu, 0·09%Ni, 0·09%Al, 0·08%Ca. It was made in the acid electric furnace under glass slag, and gave better results than several similar foreign products.

**Quality Steel for Razor Blades.** D. Tlustá and J. Votruba. (*Výzumné Práce z Oboru Železa a Oceli, State. Publ. House Tech. Lit., Prague*, 1958, 63–82). [In Czech]. Steels and methods used in razor blade production in Czechoslovakia and Sweden are discussed, together with metallographic and heat-treatment data obtained by the authors in researches on razor blade steels. Recommendations concerning changes and improvements in production technology in Czechoslovakia are made.—P. F.

**Alarm-Clock Springs.** O. Modráček and K. Staňka. (*Výzkumné Práce z Oboru Železa a Oceli, State Publ. Ho. Tech. Lit., Prague*, 1958, 139–147). [In Czech]. The development and production of alarm-clock springs in Czechoslovakia in recent times is described. Springs equal in quality to products hitherto imported are being produced.—P. F.

**Manufacture of Class II Spring Steel Wire for Motor Cycle Springs.** J. Olden and K. Werner. (*Neue Hütte*, 1958, **3**, June, 358–364). The process of manufacture of spring steel wire for motor cycle springs is followed from melting to testing. Results with the quality MK82F were compared with those from Swedish material.

**Investigation of Suitability of 18-8 (Type 304) Alloy for Superheater Service.** F. B. Snyder, T. A. McNary and F. Eberle. (*A.S.M.E.*, 1957, 57-A-174, pp. 17). Tests on non-stabilized materials in synthetic steam condensates containing 38·4 and 2000 ppm chloride showed that intercrystalline corrosion did not occur. Stress corrosion was observed with both stabilized and non-stabilized alloys and no significant difference was observed between them. A 54-month's test in service showed no intercrystalline attack.

**The Development of Silicon Transformer Steels with Anisotropic Magnetic Properties.** R. Půlpán. (*Výzkumné Práce z Oboru Železa*

*a Oceli, State Publ. Ho. Tech. Lit., Prague*, 1958, 83–90). [In Czech]. The development and production of structurally anisotropic silicon steels in Czechoslovakia are discussed.

### HISTORICAL

**Discovery of Further Smelting Furnaces in the Sieg District.** J. W. Gilles. (*Stahl Eisen*, 1958, **78**, Aug. 21, 1200–1201). Excavation of smelting furnaces near Oberstdorf in the Sieg district of Western Germany showed that one furnace was used between 1200 and 1400 A.D. Three further furnaces are of the late La-Tène period. The furnaces are described and the analysis of the slags found tabulated.

**Iron Smelting and Forging in the Mediæval Economy of the Lahne–Dill District.** R. Herwig. (*Stahl Eisen*, 1958, **78**, Sept. 18, 1332–1337). The oldest traces of iron smelting in the Lahne–Dill district of Western Germany are dated back to the La-Tène period. The history of the so-called "forest forges" and their decay in the late Mediæval Age is briefly told.—T. G.

**A Few Reminders Concerning Stainless Steels by a Great French Metallurgist: A. Portevin.** (*Aciers Fins Spec.*, 1958, July, 102–103). A note commenting on work published by Portevin leading to the discovery and concerned with the properties of stainless steels.

**Centenary Jubilee of Ilseder Hütte.** (*Stahl Eisen*, 1958, **78**, Sept. 18, 1345–1347). The history of the integrated iron and steel works is briefly reviewed.—T. G.

**The Early Ironworker and Public Opinion through the Ages.** H. R. Schubert. (*JISI*, 1959, **191**, Jan., 18–19). [This issue.]

### ECONOMICS AND STATISTICS

**Iron and Steel.** J. C. O. Harris and M. E. Palfrey. (*U.S. Bur. Mines Min. Yearbook*, 1955, **1**, 579–603). **Iron and Steel Scrap.** J. E. Larkin. (605–632). Statistics of production, import and export and utilization are given.

**The Iron and Steel Industry in Europe Trends in Economic Sectors.** Iron & Steel Committee. (*O.E.E.C. Report*, 1958, July, pp. 146). A report for the second half of 1956 and the first half of 1957. Trends in demand and supply and home and export markets are discussed and figures for production and prices are shown. Tables are appended and a raw materials forecast for 1961 is given.

**Balance Sheet of a Decade of Development of the Metallurgical Industry.** (*Met. Constr. Masini*, 1958, (6), 465–471). [In Rumanian]. A statistical review of the development of the industry in the last 10 years with statistics about investments, raw materials, imports, production trade and exports and about immediate prospects.—L. H.

### MISCELLANEOUS

**Safeguarding Steel Production.** R. M. L. Russell. (*Iron Steel Eng.*, 1958, **35**, July, 112–123). A review of the fire hazard in an iron and steel works is given, together with some general protection recommendations. The importance of automatic sprinkler installations is emphasized.

**Ventilation of Large Industrial Buildings Under Heat Load.** F. E. Tucker. (*Iron Steel Eng.*, 1958, **35**, June, 107–112). The paper outlines Weirton Steel's efforts to satisfy ventilation needs so that its employees might receive the best possible ventilation at least expense.—M. D. J. B.

**Retrospect to Three Decades of Managerial Practice.** W. Preute. (*Stahl Eisen*, 1958, **78**, July 24, 1058–1061). The managerial work of the German iron and steel industry over the last thirty years is critically reviewed. In spite of great successes a certain stagnation is now being observed. Proper training of the new generation for industrial management is emphasized.—T. G.

**Natural and Manufactured Iron Oxide Pigments.** M. L. Skow and E. V. Blankenbaker. (*U.S. Bur. Mines Min. Yearbook*, 1955, **1**, 833–839). World statistics are collected.

**$TiO_2$ Production from Native Bauxite and Red Mud.** M. Achs-Ballane. (*Kohászati Lapok*, 1958, **13**, Feb.-March, 157–158). The titania in bauxites and in bauxite sludges may be almost entirely recovered from sulphuric acid solution in presence of bivalent metals (Zn, Cu, Mg, Mn). There is only iron as impurity in the titania thus produced, which can therefore be used for production of ferro-titanium.—P. K.

**Sparking Stones of Miners' Lamps.** V. Vondrášek. (*Sbornik-Ostrava*, 1957, **3**, (3), 25–27). [In Czech]. The requirements of ignition "flints" in miners's lamps compatible with safety specifications are considered. These are met by stones containing 70–75%Ce. It is shown that metallographic and hardness tests are suitable for the rapid determination of the quality of such flints. The Fe–Ce system is discussed.—P.F.

**Prevention of Accidents in the Iron and Steel Industry.** G. Cardenas Coronado. (*First Nat. Congress Iron and Steel Industry [Mexico]*, 1955, 121–125).

**The Ventilation of Workshops in Iron and Steel Plants.** M. Hansen and H. Spitzer. (*Stahl Eisen*, 1957, **77**, Feb. 21, 204–215). A report from a Committee of the Power and Fuels Dept. of the V.D.E. Calculations and design factors are thoroughly treated.

**Glossary of Technical Terms Applicable to the Iron and Steel Industry.** (*Second National Congress of the Iron and Steel Industry, [Mexico]*, 1957, 297–308). A wide range of terms, with definitions (all in Spanish) is given.

**Modern Computer Analysis for the Design of Steel Mill Control Systems.** J. E. Reider and P. Spergel. (*Trans. A.I.E.E.*, 1957, **76**, Part 2, 105–110).

**Some Observations on the Management of American Steel Plants.** *A.* **Working Efficiency.** W. Schütte. (*Stahl Eisen*, 1958, **78**, June 26, 873–875). The author reports on impressions gained during a visit to a number of American steelworks, in particular rolling mills. *B.* **Differences in Organization of the Plants Between German and American Practice.** H. Spitzer. (876–882). The author analyses and discusses the findings of the Harbison Committee on the management of American steel works and compares the structures of organization of German and American plants. The high efficiency of the American plants as compared with German works is stressed.

## *BOOK NOTICES*

HARNED, H. S., and B. B. OWEN, "*The Physical Chemistry of Electrolyte Solutions*." Third edition. A.C.S. Monograph No. 137. 8¾ × 6 in., pp. xxxii + 803. Illustrated. New York, London, 1958: Reinhold Publishing Corporation, Chapman & Hall Ltd. (Price £8)

This book first appeared in 1943 and a second edition was needed by 1949. Revision in the second edition was achieved by the addition of an appendix, but the drastic revision for the present edition has necessitated a 25% increase in the number of pages, accompanied by a 50% increase in price. The latest revision has every appearance of being thorough; it contains several references to work published in 1957 and a considerable number for earlier years in this decade.

One of the major tasks involved in this revision was that of recalculating the tables in Chapter 5 and the numerical coefficients of all numbered equations throughout the book to conform with the values recommended by the Sub-committee on Fundamental Constants, National Research Council, 1952.

New material includes a section on Irreversible Thermodynamics in Chapter 1, an outline of Mayer's cluster-sum theory of thermodynamic properties in Chapter 3, the theory of Onsager and Kim for the conductivity of solutions containing s-ions and their theory of the Wien effect for strong electrolytes of any valency type in Chapter 4, recent work on the self-diffusion of ions in concentrated solutions in Chapter 6, ion pair formation as a function of dielectric constant and measurement of the Wien effect by a differential pulse transformer bridge in Chapter 7, and a modernized version and expansion of Bjerrum's theory of ionic hydration by Stokes and Robinson in Chapter 12.

Some examples of careless proof reading have been noted. The word "quarternary" appears frequently, e.g. on p. 271, and the spellings "conductomatric" (p. 245) and "intergrating" (p. 280) occur. References seem to have been omitted from Table 13-1-2A on p. 740.—F. W. SALT.

HARWOOD, J. J., *et al.*, "*The Effects of Radiation on Materials*." 9 × 6 in., pp. 355. Illustrated. New York, London, 1958: Reinhold Publishing Corporation, Chapman & Hall Ltd. (Price 84*s*.)

The changes produced in materials by atomic radiation have been studied intensely since nuclear reactors were introduced. Although the subject is still in its infancy, the rapid development of nuclear energy has quickly brought it to prominence. As the damage caused by atomic projectiles is itself atomic in character, atoms being displaced and vacancies, interstitial atoms and impurities introduced, an understanding of the behaviour of materials during radiation calls for an atomic approach. Metals are fortunately less susceptible to radiation damage than most other materials, mainly because the displacement of their electrons produces little change in properties. However some physical properties, e.g. internal friction, are radically altered after a few minutes in a nuclear reactor, and after long times many properties are changed, and some metals, e.g. uranium, become almost unrecognisable.

People interested in materials should keep abreast of this rather exciting subject so that they can advise on the best materials to use, and produce better materials for use in reactors than exist at the moment. The subject is sufficiently new for there to be very few review articles to assist, and this book is a very welcome addition. It is really a collection of papers which were delivered at a Colloquium of the same title, jointly sponsored by the Office of Naval Research and the Martin Company at Johns Hopkins University, Baltimore, in March 1957. This method of compiling a book has the advantage that it can be speedily written, and that each subject receives specialist treatment, but suffers from overlap and non-uniformity. The first chapter, or paper, by Dienes introduces the subject in terms of the elementary atomic concepts and is a useful starting point for the chapters which follow. These chapters cover effects in metals and alloys (Billington, Simnad), semiconductors (Fan and Lark-Horovitz), inorganic dielectric materials (Smoluchowski), organic substances (Burton) and polymers (Charlesby, Restaino). Two further chapters discuss the behaviour of various reactor components, cores, coolants, control rods (Weber), moderators, shielding and auxiliary equipment (Hennig). The experimental techniques are described separately (Wilson) and bring home some of the difficulties involved. A sectionalized bibliography of 779 references is included at the end of the book in addition to the references associated with each chapter.

Apart from some uncut pages in the copy reviewed and a photograph on page 28 which has been badly reproduced the general standard of printing and binding is good. This book should be very useful to the specialist who is not dismayed by the confusion of a subject which is too new to be tidy.—R. S. BARNES.

HARWOOD, JULIUS J. (Editor). "*The Metal Molybdenum.*" 9 × 6½ in., pp. 696. Illustrated. Cleveland, Ohio, 1958: American Society for Metals. (Price $12.50)

ROSTOKER, WILLIAM. "*The Metallurgy of Vanadium.*" 9 × 6 in., pp. 185. Illustrated. New York, London, 1958; John Wiley & Sons, Inc., Chapman & Hall, Ltd. (Price £3 8*s.*)

The origin of the first of these books was a symposium "The Technology of Molybdenum and Its Alloys" conducted by the (American) Office of Naval Research in September 1958: to the papers delivered there, several more have been added to provide a more complete coverage of the subject. The book therefore contains an immense amount of valuable material, ranging from the arc-melting of molybdenum to the determination of gases in the metal, and from its brazing to its metallography.

The second book is equally thorough, and works steadily on from the extractive metallurgy of vanadium through its properties to its metallography, and finishes with its uses as an alloy addition. There is no original material therein, but the sources of information are widespread: many of them are non-American.—J. P. S.

"*Nachschlagewerk Stahlschlüssel.*" 4. Auflage. 12 × 9 in., pp. 96. Marbach A/N, 1958: Verlag-Stahlschlüssel. (Price DM.30.–)

This volume, whose title might be translated as "Encyclopaedic Skeleton Key to Steel," is wholly composed of tables whereby one may not only obtain the composition, mechanical properties, and appropriate heat-treatment of German steels—arranged in groups, constructional steels, heat-resisting steels, high-speed steels and so on—but also determine the trade-name of the equivalents made by a number of German firms. The same treatment has been applied to a number of "*hart-metalle,*" and some foreign equivalents are attached.

J. P. SAVILLE.

OTTO, C. A. (Ed.). "*Electric Furnaces.*" 8⅛ × 5½ in., pp. vii + 248. Illustrated. London, 1958: George Newnes Ltd. (Price 35*s.*).

This is a composite work edited by C. A. Otto who has also written three chapters, the other eleven chapters being contributed by specialists in different fields. A book of this kind necessarily lacks a certain continuity and there is inevitable repetition. This is to a large extent due to the fact that the subject matter is sub-divided into sections or chapters, some dealing with types of furnaces, others with applications, whilst the last two chapters are concerned with control gear and refractories.

There is an important omission in the book in so far as electric smelting furnaces are not dealt with at all. The author of Chapter 2 says that such furnaces are of little interest in this country due to the high cost of power. In fact, smelting furnaces of a total electrical demand of over 120 000 kVA are in use or building in this country, and this represents considerably more load than many of the other electric furnace applications. Chapter 3 is a competently written section by R. J. Sarjant on melting furnaces and all the various types are dealt with in the compass of some thirty-five pages. The following sections on resistance furnaces and heat treatment overlap somewhat, but they give the reader a good survey of this field. Induction heating and its applications are well described in Chapter 6, with examples of operating practice. Then the subject matter turns to vacuum furnaces, dealing mainly with vacuum melting, but heat treatment and brazing are also mentioned. The ensuing section deals with salt bath furnaces, including their use for the heat treatment of steels, but this is followed by chapters covering the glass industry, vitreous enamelling, pottery and plastics, which are not of direct metallurgical interest.

Covering such a wide field, the book can only treat the subject on broad lines and must therefore be regarded as a general review. Moreover, the reader is not given any references or bibliography to enable him to supplement the information. It is well produced, with excellent illustrations, but these are numbered consecutively only within each chapter, which tends to emphasise the sectional nature of the work.—A. G. ROBIETTE.

GUNTHER, RUDOLF. "*Glass-Melting Tank Furnaces.*" Translated from the German by John Currie with a foreword by Emeritus Professor W. E. S. Turner. 8½ × 5½ in., pp. xiv + 232. Illustrated. Sheffield, 1958: Society of Glass Technology. (Price 60*s.*).

This is a most excellent work on a specialized subject ostensibly not of direct interest to the steel industry. On the other hand, many of the problems encountered in melting glass are not dissimilar from those met with in open-hearth practice.

The book contains chapters dealing with refractory construction, recuperation and regeneration of heat, fuel efficiency, and there is an excellent chapter on fundamental thermal calculations. It is both detailed and practical and could be most usefully studied by all those interested in the design and operation of steel melting furnaces.

The work is well documented with tables and illustrations, and the translator and his collaborators have done a commendable job in the conversions from the German original. The production is extremely good, and the references and bibliography are comprehensive.—A. G. ROBIETTE.

POUNDS, NORMAN J. G., and WILLIAM N. PARKER. "*Coal and Steel in Western Europe.*" 9 × 6½ in., pp. 381. Illustrated. London, 1958: Faber and Faber Ltd. (Price 45*s.*).

This country is excluded from these authors' scope, because it stands apart from the European Coal and Steel Community with whose members they are most concerned. It enters into view only in so far as British inventions and technique influenced developments on the Continent: therefore, they refer only indirectly to the influence of Continental development on British practice and economic outlook.

The important sections of the book deal with the growth of coal mining and iron- and steel-making in France, Germany, and Belgium from about the 17th century to the present day, noting at all stages the effects of political changes as well as those of new techniques: for instance whereas in the 18th century the French iron-melting industry was hampered by lack of capital and by government control, that of Prussia was fostered by the enlightened State. Yet, in spite of this early difference, the great step towards the creation of an iron and steel industry composed of a few large units, the movement of iron-making to the coal-fields, took place at much the same period in both countries when the coming of railway transport made it possible.

The Belgian industry, on the other hand, which also grew to considerable size in the same period, was the result of a much more deliberate act of creation, and the careful imitation of English models; native ore and coking coal seemed abundant, and the rate of expansion was unparalleled. Nevertheless, by the end of the 19th century, Belgium had become dependent upon foreign ores and coking coal.

The authors deal with many factors of this kind; observing, for instance, that the deep coal of the Ruhr could not be reached until a metallurgical industry had developed, and that the only large ore deposit in France, the *minette* of Lorraine, could be worked only by the basic process.

In conclusion, they examine the purpose of the European Coal and Steel Community, and express the hope that it will bring about conditions in which "gradual expansion, leading perhaps to eventual stability," can be achieved without serious cyclic disturbance.—J. P. SAVILLE.

ROE, LAWRENCE A. "*Iron Ore Beneficiation.*" 8½ × 5½ in., pp. ix + 305. Illustrated. Lake Bluff, Illinois, 1957: Minerals Publishing Company. (Price 45*s.*).

The subject of iron ore beneficiation has received a great deal of attention for many years and especially in the last decade or so. Shortages of high-grade ore, sometimes local and temporary, sometimes long-term, have stimulated much research into economic methods of utilizing low-grade ores. This book attempts to collect together, it is believed for the first time, information which has been published over many years and in many parts of the world. The emphasis, however, is perhaps inevitably more on experience and practice in America than elsewhere. The first two chapters deal with the history of iron-ore beneficiation and with sources of iron ore and economic factors. A chapter is devoted to a full discussion of the iron ore minerals. This is followed by five chapters dealing with the various physical methods used for concentrating iron ores, including washing and jigging, heavy-media processes, tabling and spiral treatment, flotation and magnetic separation. These chapters are illustrated with flowsheets and results from plant practices. One is reminded by the chapters on pyrometallurgical processing, hydrometallurgy and agglomeration that the term beneficiation is not confined to the physical concentration of iron ores.

A notable feature of this book is the very comprehensive list of references which are referred to in the text, as well as additional references at the end of each chapter.

With the present rate of progress in this field, such a book will need frequent revision to bring it up to date. Nevertheless, for all those engaged in the concentration and preparation of ores prior to smelting this book will be invaluable as a reference book and as a starting point for the study of many specialized aspects of iron ore beneficiation. The author is to be congratulated for his clear and concise treatment of a very broad subject.—G. E. DAVIES.

"*Schweissen von Gusseisen.*" Fachbuchreihe "Schweisstechnik." Band 4. 9 × 6½ in., pp. viii + 116. Illustrated. Düsseldorf; Braunschweig, 1958: Deutscher Verlag für Schweisstechnik (DVS) GMBH; Friedrich Vieweg & Sohn. (Price DM. 22.80).

This book, "Welding of Cast Iron," which is the 14th of a general series published by the German Institute of Welding, contains chapters by five continental experts and provides a comprehensive review of the metallurgical, mechanical, practical and economic problems involved in welding commercially available alloyed and unalloyed grades of cast iron, with specific references to German material specifications. An expert and concise approach to the academic and practical aspects of the subject is ensured by the high technical qualifications and wide experience of the contributors.

The first chapter deals with the metallurgy of all types of cast iron including the inoculated and alloy grades as well as the theoretical considerations necessary for an understanding of the associated welding problems. The fusion welding processes involving a high degree of preheat are discussed in the second chapter which deals with such topics as the choice of filler materials, welding procedures, weld preparations, heat treatment, and the associated properties of the weldment. The practical aspects of fusion welding are illustrated by fifteen well chosen examples with a useful evaluation of the material and fuel consumption and the time associated with various operations. The problems of "cold welding" by the metal-arc process and the brazing of cast irons are considered in Chapter 3. This chapter also includes a discussion of the applicability of nickel and Monel-base electrodes in different combinations for welding cast iron. The two last chapters of the book deal with welding cast

iron to steel and contain a comprehensive survey of the weldability problems associated with different grades of malleable cast iron. The recommendations and technical details contained in the book are endorsed by the major welding manufacturers and welding establishments in Germany and Switzerland.

The book can be recommended to anyone with a knowledge of German who is interested in any aspect of the welding of cast iron. One major disadvantage to the British reader, however, is the cross referencing required to the equivalent grades and specifications of cast irons produced in Germany and in this country.—M. C. T. Bystram.

Siebel, Erich. "*Handbuch der Werkstoffprüfung.*" 2. Auflage. Erster Band: "*Prüf- und Messeinrichtungen.*" Herausgegeben von Prof. Dr.-Ing. E. Siebel und Dipl.-Ing. N. Ludwig. pp. xvi + 890. Illustrated. Berlin, 1958: Springer-Verlag. (Price DM. 148.50).

This weighty and impressive volume is one of a series of five on Material Testing. (Volume II on Testing of Metallic Materials, whose second edition appeared in 1955, is the only other one of interest to metallurgists). It is a revised edition, and consists of twelve sections contributed by different authors. The approach adopted is down to earth, and the book consists largely of descriptions of apparatus with guidance on the techniques of its use.

The first nine sections deal with apparatus and techniques for mechanical testing, hardness testing, strain measurement, and non-destructive testing. Here the treatment is admirably full, not to say exhaustive, except that the machines specifically described are almost exclusively of German manufacture; this may be of some disadvantage to British users but the main value of the book lies in the more general discussions of principles and techniques. The last three sections deal with metallography, chemical analysis and spectrographic analysis. These, and in particular that on chemical analysis, are not of a standard to interest the specialist and so fall below the level of the previous sections.

The book is beautifully produced and is illustrated with more than 1,000 clear diagrams and photographs. It would undoubtedly be an extremely useful reference book in laboratories engaged in material testing, but at the price most libraries will have to think twice before buying it. Professor Siebel might have done better to have cut the price by omitting material more fully treated in other recent text books, but on the whole he and the authors have done a monumental piece of work.

J. D. Baird.

# *NEW PUBLICATIONS*

"*ABC der Stahlkorrosion unter besonderer Berücksichtigung der chemisch beständigen Stähle.*" 4to, pp. 498. Illustrated. Düsseldorf, 1958: Mannesmann Verkaufsgemeinschaft.

Aitchison, L., and W. I. Pumphrey. "*Using Steel Wisely.*" 8½ × 5½ in., pp. xxii + 241. Illustrated. London, 1958: Macdonald & Evans, Ltd. (Price £2 4*s.*).

Allendorf, Hans. "*Präzisionegiessverfahren mit Ausschmelzmodellen.*" 2 Auflage. 8vo, pp. viii + 326. Illustrated. Leipzig, 1958: Fachbuchverlag. (Price DM. 18.–).

American Society for Metals. "*Metals for Supersonic Aircraft and Missiles.*" Proceedings of the Conference: Heat Tolerant Metals for Aerodynamic Applications, January 28, 29, 1957, University of New Mexico, Albuquerque. 9 × 6 in., pp. vi + 432. Illustrated. Cleveland, Ohio, 1958: The Society. (Price $7.50).

American Society for Testing Materials. "*Symposium on Brittle Failure of Rotor Forgings.*" STP. No. 231. 9 × 6 in., pp. v + 43. Illustrated. Philadelphia, 1958: The Society. (Price, Members, $1.60; non-Members, $2.00).

American Society for Testing Materials. "*Symposium on Determination of Gases in Metals.*" 9 × 6 in., pp. ii + 60. Illustrated. Philadelphia, 1958: The Society. (Price, Members, $1.80; non-Members, $2.25).

American Society for Testing Materials. "*Symposium on Spectrochemical Analysis for Trace Elements.*" STP. No. 221. 9 × 6 in., pp. v + 79. Illustrated. Philadelphia, 1958: The Society. (Price, Members, $2.20; non-Members, $2.75).

American Society for Testing Materials. "1957 *Supplement to the Metal Cleaning Bibliographical Abstracts.*" Prepared by Jay C. Harris. STP No. 90–D. 9 × 6 in., pp. 40. 1958, Philadelphia: The Society. (Price, Members, $1.50; non-Members, $2.00).

"*Aufgaben deutscher Forschung. Band 3. Technik.*" 2 Auflage. Im Auftr. der Ministerpräsidenten Fritz Steinhoff zagestellt u. hrsg. von Leo Brandt. 8vo, pp. xxi + 824. Köln u.s.w., 1958: Westdeutscher Verlag. (Price DM. 56.40).

British Standards Institution. B.S. 1333: 1958. "*Acid Resisting Silicon Iron Pipes and Pipe Fittings.*" 8¼ × 5¾ in., pp. 23. Illustrated. London, 1958: The Institution. (Price 6*s.*).

British Standards Institution. B.S. 1633: 1958. "*Steel for Land Boilers, Receivers.*" 8¼ × 5¾ in., pp. 22. Illustrated. London, 1958: The Institution. (Price 6*s.*).

British Standards Institution. B.S. 3014: 1958. "*Welded Austenitic Stainless Steel Tubes.*" 8¼ × 5¾ in., pp. 15. London, 1958: The Institution. (Price 4*s.* 6*d.*).

British Standards Institution. B.S. 3029: 1958. "*Determination of Compressibility of Metal Powders.*" 8¼ × 5¾ in., pp. 6 + ii. Illustrated. London, 1958: The Institution. (Price 3*s.*).

British Standards Institution. B.S.T. 55 Amendment. May 1958. "35 *Ton Chromium Nickel Non-Corrodible Steel Tubes (Suitable for Pipe Lines) for Aircraft Purposes.*" 13 × 8¼ in., pp. 2. London, 1948, 1951, 1958: The Institution. (Price 2*s.* 6*d.*).

"*Compressors and Vacuum Pumps and their Lubrication.*" 9½ × 6½ in., pp. xiii + 225. Illustrated. London, 1958: The Shell Petroleum Co. Ltd.

Deutschen Verbandes für Schweisstechnik. "*Steigerung des Wirtschaftlichkeit und der Produktivität durch die Schweisstechnik.*" (Vorträge der grossen schweisstechnischen Jubiläumstagung des D. Verb. für Schweisst. Essen, June 25–28, 1957). 4to, pp. iv + 143. Illustrated. Braunschweig, 1957: Vieweg. (Price DM. 18.50).

Fuchs, Walter and Hans Weyerstrass. "*Entwicklung eines Heissfilters zur Reinigung von Gichtgas eines mit Kohle betriebenen Niederschachtofens.*" Forschungsberichte des Wirtschafts- und Verkehrsministeriums Nordrhein-Westfalen. Nr 400. 4to, pp. 75. Illustrated. Köln und Opladen, 1958: Westdeutscher Verlag. (Price DM. 20.20).

"*Giesserei Lexicon.*" Herausgegeben von A. Schulenburg. 9½ × 6½ in., pp. xiv + 858. Illustrated. Berlin, 1958: Fachverlag Schiel & Schön GmbH. (Price DM. 58.–).

Golikov, I. N. "*Dendritnaia likvaziya v Stali.*" 8vo, pp. 206. Illustrated. Moscow, 1958: Gossudarstvennoe nauknotechnytachesskoi Isdatesstvo Literatury po Chernoi i Zvetnoi Metallurgii. (Price DM. 4.60).

Institut National des Sciences et Techniques Nucléaires. "*Symposium de Métallurgie Spéciale.*" Saclay, les 27 et 28 Juin 1957. 10½ × 8¼ in., pp. 128. Illustrated. Paris, 1958: Presses Universitaires de France.

The Institution of Mechanical Engineers. "*Proceedings of the Conference on the Properties of Materials at High Rates of Strain. London, 30 April–2 May* 1957." 11 × 8½ in., pp. viii + 268. Illustrated. London, 1958: The Institution. (Price £3 3*s.*).

"*Inventaire des Périodiques Soviétiques reçus en France par les Bibliothèques et les Organismes de Documentation en* 1957." 10½ × 8 in., pp. 84 + 5. Paris, 1958: Direction des Biblioteques de France (I.P.P.E.C.).

IRSID. "*Coulée de l'Acier. Les Cahiers du CESSID. XV.*" 11½ × 9½ in., Cahier I, pp. 55, 49, 52; Cahier II, pp. 40, 32, 24, 62; Illustrated. Metz, 1958: CESSID. (Price 6250 fr.).

Kelf-Cohen, R. "*Nationalisation in Britain. The End of a Dogma.*" 8½ × 5½ in., pp. x + 310. London, 1958: Macmillan & Co. Ltd. (Price 25*s.*).

Lehfeldt, Wilhelm. "*Ultraschall in der Feinbearbeitung.*" (Schriftenreihe Feinbearbeitung, Heft 26). 8vo, pp. 88. Illustrated. Stüttgart, 1958: Deva-Fachverlag. (Price DM. 5.80).

Merriman, A. D. "*A Dictionary of Metallurgy.*" 11 × 8½ in., pp. xv + 401. Illustrated. London, 1958: Macdonald & Evans Ltd. (Price £6 6*s.*).

Piwowarsky, Eugen. "*Hochwertiges Gusseisen (Grauguss) seine Eigenschaften und die physikalische Metallurgie seiner Herstellung.*" Zweite Auflage Neudruck. 9¾ × 6¾ in., pp. xii + 1070. Illustrated. Berlin, 1958: Springer-Verlag. (Price DM. 135.–).

Roper, Francis D., "*Cores and Core-Making.*" 8½ × 5½ in., pp. 125. Illustrated. London, 1958: George Allen & Unwin Ltd. (Price 21*s.*).

Schmidt, Hans, and Herbert Dickmann. "*Bronze-und Eisenguss. Bilder aus dem Werden der Giesstechnik.*" (Ein Bericht über die Historische Sonderschau der Internationalen Giessereifachmesse 1956). 4to, pp. 121. Illustrated. Düsseldorf, 1958: Verein Deutscher Giessereifachleute. (Price DM. 20.–).

Schneider-Bürger, M. "*Stahlbau-Profile.*" Herausg. vom V.d.E. in Zusammenarb. mit der Beratungsstelle f. Stahlverwendung. 9 Auflage. 8¼ × 5¾ in., pp. 48. Illustrated. Düsseldorf, 1958: Verlag Stahleisen m.b.H. (Price DM. 3.50).

"*Seibt Export Directory of German Industries.*" Editor: Dr. A. Seibt. 14th ed. 11½ × 7¾ in., pp. 1215 + 310. Illustrated. Lengdorf and München, 1958: Seibt-Verlag. (Price £3 5*s.*).

"*Textile Machinery Index.*" 2nd ed. 10½ × 6½ in., pp. 682. Illustrated. Oldham, 1958: John Worrall Ltd. (Price £2 2*s.*).

"*Transformation Characteristics of Direct-Hardening Nickel-Alloy Steels.*" 3rd edition. 9½ × 6½ in., pp. 91. Illustrated. London, 1958: The Mond Nickel Company Ltd.

United Nations. "*Problems of the Steel Making and Transforming Industries in Latin America.*" Vol. 1. Report of the Latin American Meeting of Experts on Steel Making and Transforming Industries, Sao Paulo, October 15–28, 1956. pp. 62 [in English]. Vol. 2. Siderurgia, pp. 258. Illustrated [in Spanish]. 11 × 8½ in. Vol. 1, New York, 1958 (Price 5*s.*). Vol. 2, Mexico, 1957 (Price 18*s.*).

Wellinger, Karl, and Gimmel Paul. "*Werkstoff-Tabellen der Metalle. Bezeichnung, Festigkeitswerte, Verwendung, Lieferwerke.*" 5. Auflag. 8vo, pp. 305. Stuttgart, 1958: Kröner. (Price, DM. 18.–).

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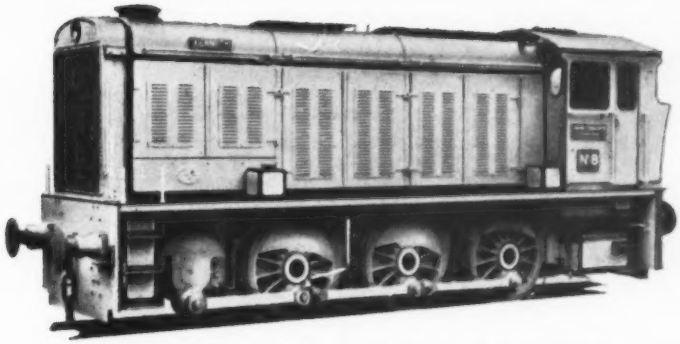

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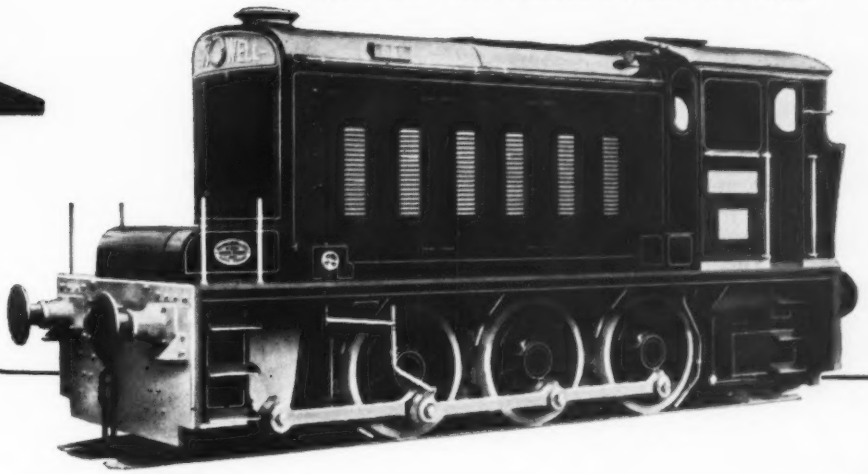

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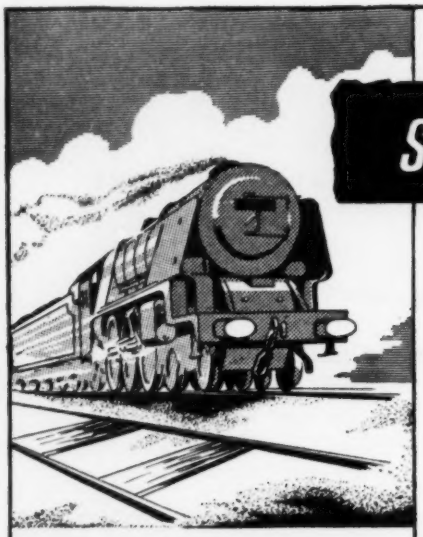

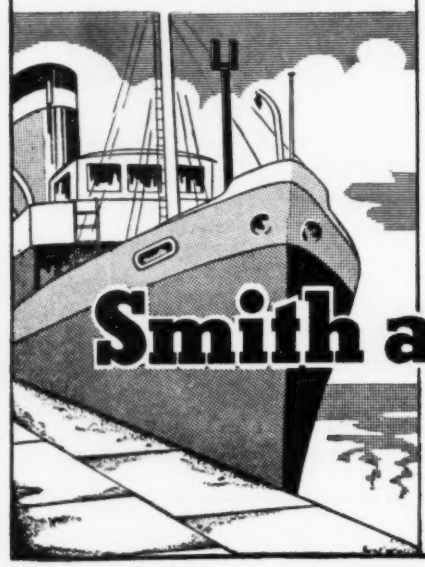

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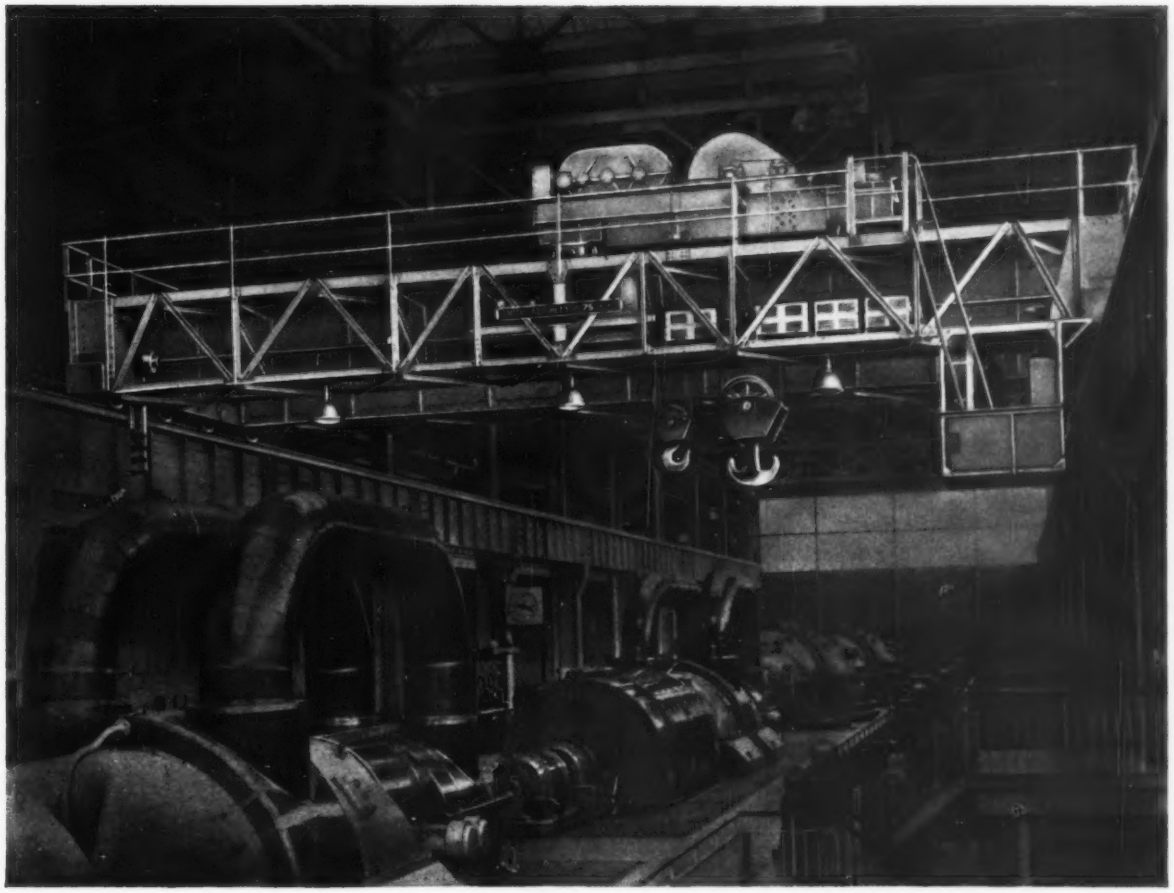

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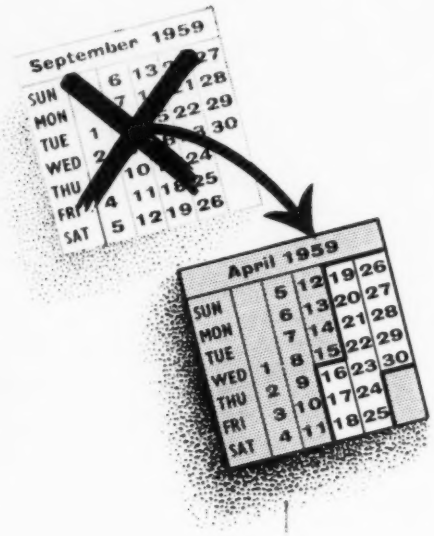


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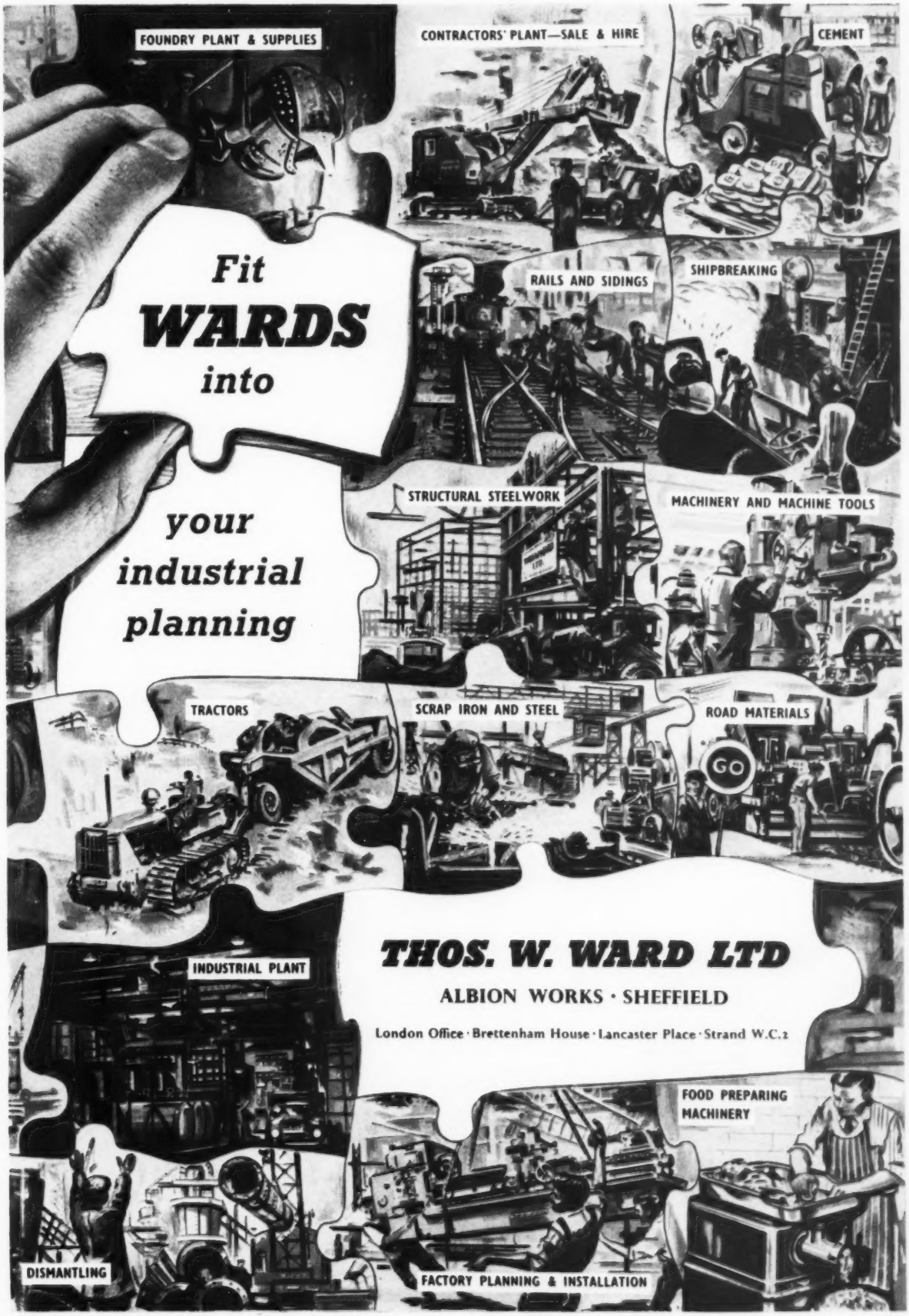
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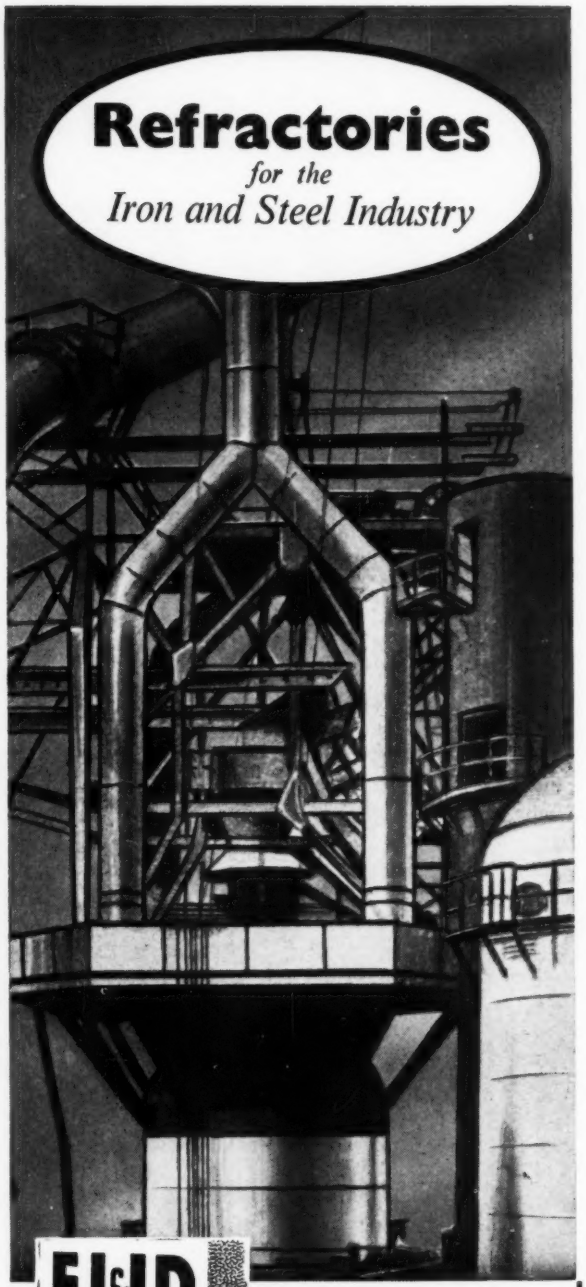


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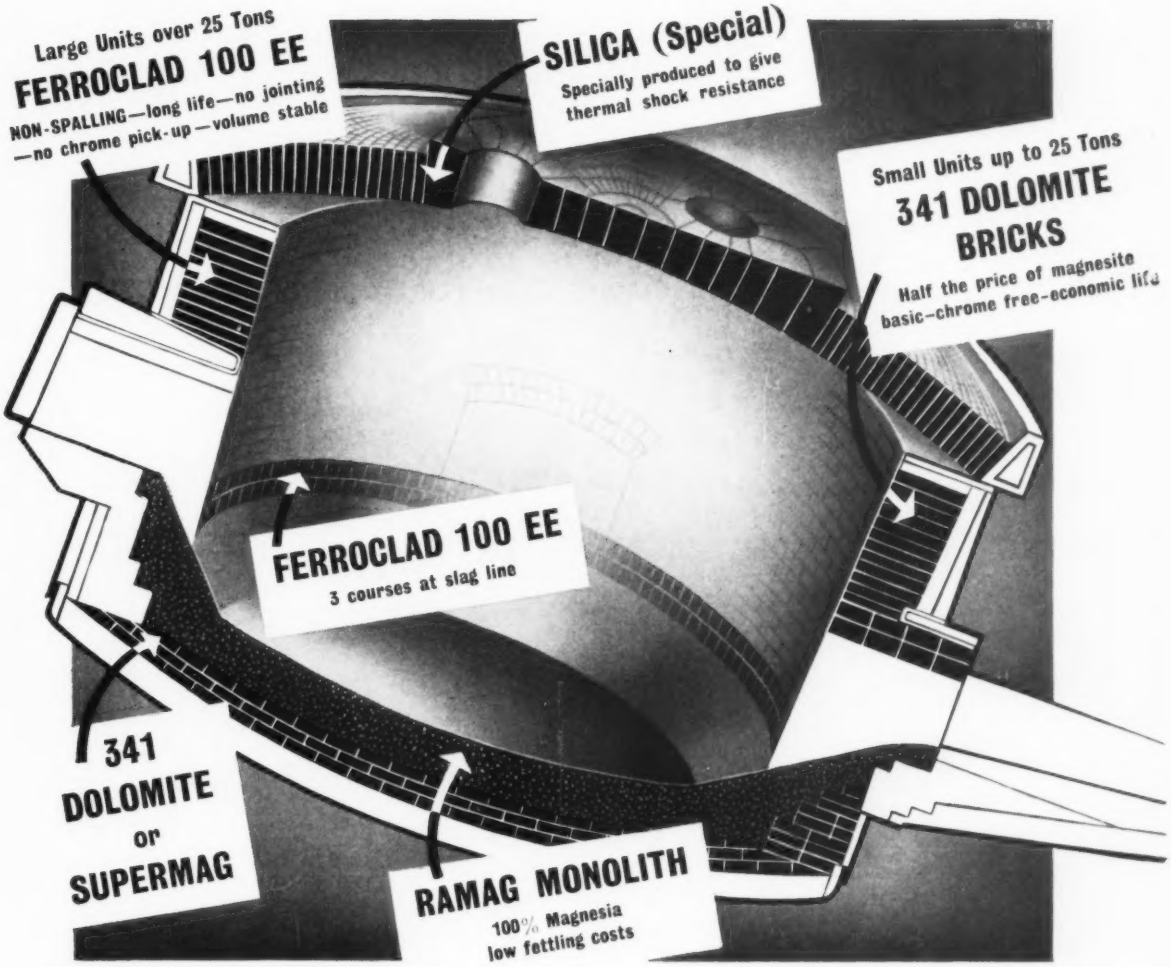


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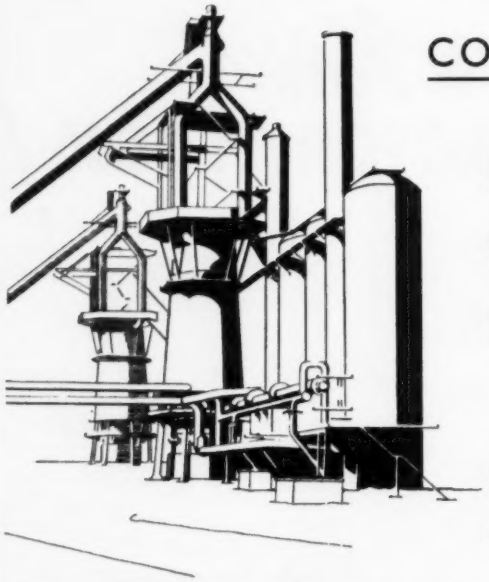

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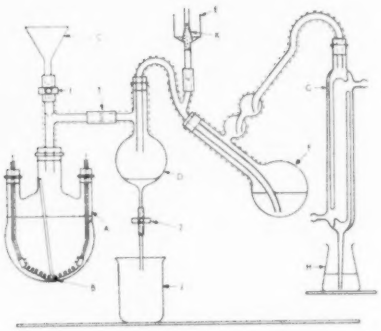

Great interest has always been shown in the role of nitrogen in determining the properties of ferrous alloys. Reliable and accurate methods are therefore necessary for determining the content of this element in alloys.

The establishment of the reproducibility and accuracy of analytical methods to serve as routine or reference procedures for the many alloys now produced is a major undertaking. It was for this reason that the Nitrogen Group of the Steelmaking Division of the British Iron and Steel Research Association was set up eight years ago. The Report of the Group has now been published as No. 62 in The Iron and Steel Institute's Special Report Series. The information it contains will be of inestimable value to all those interested in this subject.

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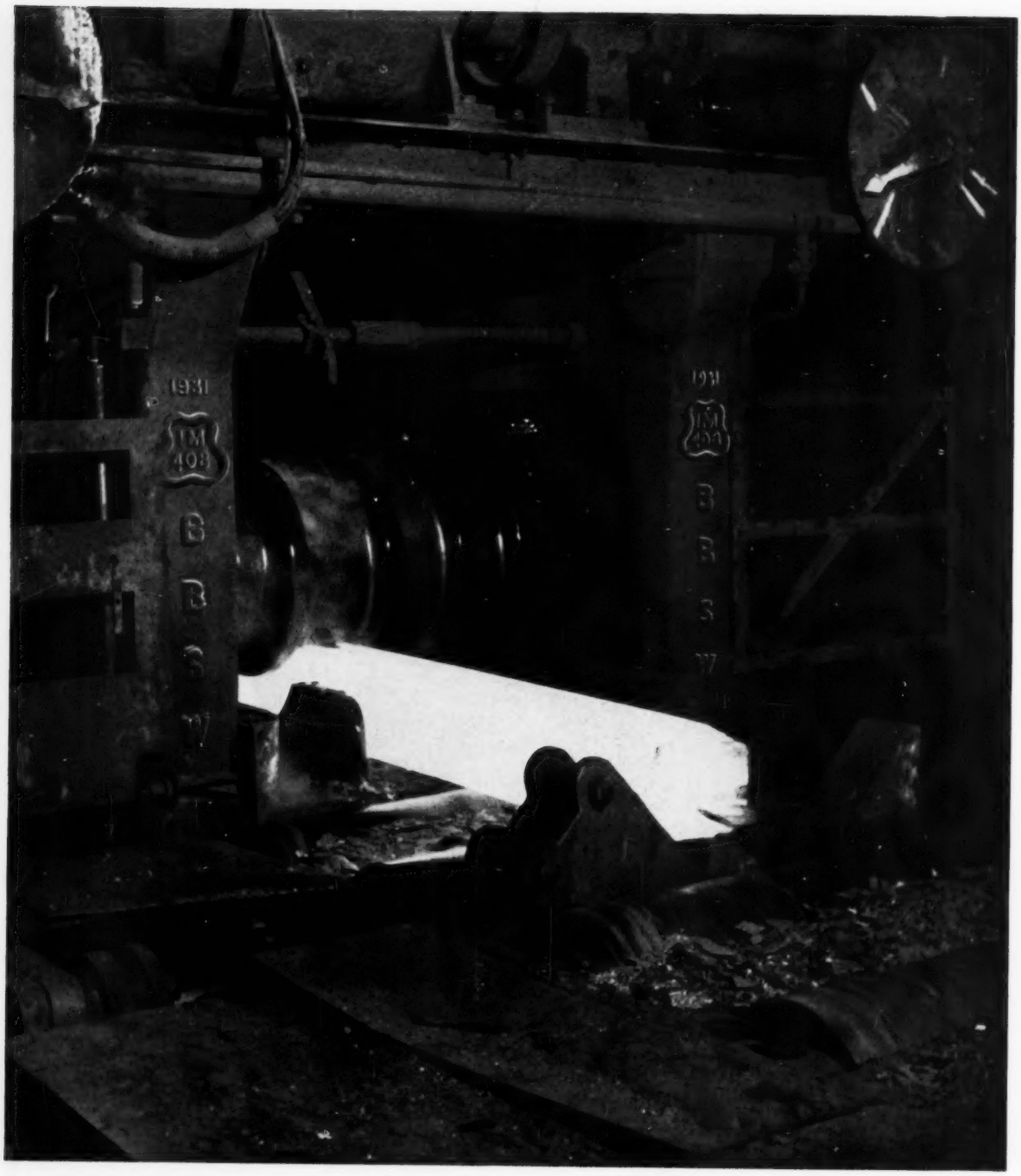


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